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CASTING PROPERTIES OF HEAT-RESISTING ALLOYS

- USSR -

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/Following is the translation of three articles from the Russian-language publication Liteynyye Svoystva Zharoprochnykh Splavov (English version above) (Transactions of the Leningrad Polytechnic Institute imeni M.I. Kalinin), No 224, Moscow, Metallurgizdat, 1963. Complete bibliographic information accompanies each article./

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PRIMARY CRYSTALLIZATION OF HEAT-RESISTING ALLOYS

[Following is the translation of an article by A.Ya. Ioffe in the Russian-language publication Trudy LPI (Transactions of Leningrad Poly. Inst.), No 224, Moscow, 1963, pp 97-112.]

Primary crystallization has a decisive influence on the properties of castings from the heat-resistant alloys since in the majority of the cases recrystallization of these alloys in the solid state does not take place. The present paper reports on a study of the iron and nickle base alloys having a constant chromium content. In addition, the alloys are alloyed with various strengthening elements (see Yu.A. Nekhendzi, The Selection of Heat-Resistant Alloys for the Study of Casting Properties, page 9 of present collection).

In connection with the fact that our reference alloy is an iron-nickel-chromium alloy, we are interested in the composition of the structural phases which are formed in this alloy under equilibrium condition (Figs 1,6).

The state diagram of the nickel-chromium alloys shows that this system is of the eutectic type with mutual limited solubility. The nickel-base solid solutions have a γ structure, those with chromium base are α while the intermediates are $\alpha + \gamma$. The limit of the solubility of chromium in nickel at normal temperature amounts to 33% (atomic), i.e. about 30% (weight). At temperatures above 700C this limit increases and at the temperature of the eutectic transformation (1340C) reaches 50% (atomic). Similarly the solubility of nickel in chromium varies from 11% (atomic), i.e. 9% (weight), at normal temperature to 30% at 1340C. The eutectic mixture consists of 45% (atomic) Ni and 55% (atomic) Cr (50% by weight of each element). At 550C in the region of the solid solution of the chromium in the nickel there is formed the chemical compound Ni_3Cr. As a result of the low

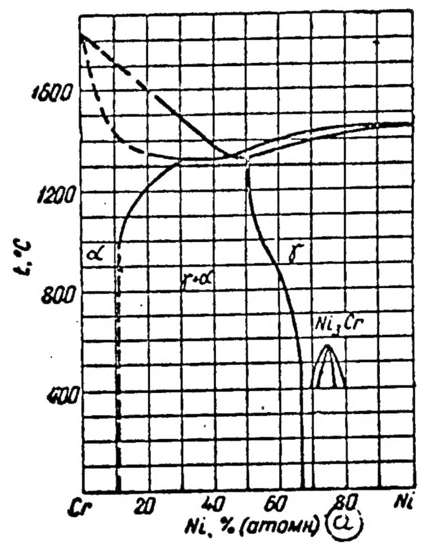

Fig 1. State diagram of the Ni-Cr system alloys

a - atomic %

temperature the rate of formation of this compound is not high.

Depending on the composition, the ternary alloys Fe-Ni-Cr (see Fig 6) (Refs 3,4) may have differing structure: ferritic α, austenitic γ and mixed $\alpha + \gamma$. In the process of slow cooling from 950C, in the solid state there may be formed the intermetallic σ-phase. The position of the boundaries of the structural regions α and γ varies as a function of the temperature: with an increase of the temperature the boundary of the γ region is shifted toward the lower nickel content. A purely austenitic structure containing 12% Cr and 12% Ni can be obtained at normal temperature in the alloy. With an increase or decrease of the chromium content a larger nickel content is required to obtain the austenitic structure. Thus, in the alloys studied containing about 20% Cr the minimum nickel content for an austenitic structure is 15%. If the alloy contains more than 10% Cr the austenitic structure can be obtained only with a nickel content equal to or greater than the critical value (Refs 5,6).

However, the state diagrams considered for the Ni-Cr and Fe-Ni-Cr alloys cannot completely characterize the structure and the phase composition for nonequilibrium cooling, i.e. in the actual conditions of the production of

castings with variable cooling rate, different liquid state of the metal depending on the melt conditions, differing contamination by nonmetallic inclusions, etc.

The state diagrams presented cannot characterize the structures found for the alloys tested for the additional reason that these alloys were alloyed with those elements which expand both the region of the γ solid solution (manganese, nitrogen) and the region of the σ solid solution (silicon, molybdenum, tungsten, niobium, titanium). In any case, as mentioned above (Nekhendzi), the alloys studied having a constant hydrogen (0.12 and 0.35%) and chromium (about 20%) content had such a balanced composition that they were in essence austenitic with a hardening carbide, carbonitride, or intermetallic phase.

Depending on the conditions, the primary crystallization of the alloys with both the iron and nickel bases leads to some growth of the zones of the austenitic equiaxed and the columnar crystallites, to some particular dimension of the primary grains of the primary solid solution and the hardening phase. The sensitivity of the alloys to a change of the crystallization conditions (for example cooling rate, superheat) differs and depends on the nature of the alloys. Therefore the study of the primary crystallization of the alloys mentioned above was carried out in conditions where the primary factors affecting the nature of the crystallization were varied over a wide range.

1. METHOD OF INVESTIGATION

The primary structure is characterized by the macrostructure, the dimensions of the primary grain, and the variation of the shape of the grain across the section. As we mentioned, these parameters depend not only on the alloy composition but also on its liquid state and the rate of cooling in the solidification period. In the present study the variation of the cooling rate was achieved by variation of the section of the samples, the temperature of the form and the temperature of the metal during casting. The effect of the liquid state was studied only in the most important cases by variation of the metal superheat temperature during melting and pouring.

In connection with the fact that the change of the rate of cooling, no matter how it is accomplished, in the final analysis affects the duration of the solidification (Refs 9,13), the factors listed above were evaluated and

analyzed on the basis of the results of the measurement of the duration of the solidification of standard specimens (see V.Ya. Bilyk, Critical Temperatures and the Duration of Solidification of the Heat-Resistant Alloys, present collection, page 61).

Two values were determined to characterize the macrostructure of the alloys studied: the grain size (area) and the width of the transcrystallization zone, i.e. the zone of the columnar crystals with relation to the entire section.

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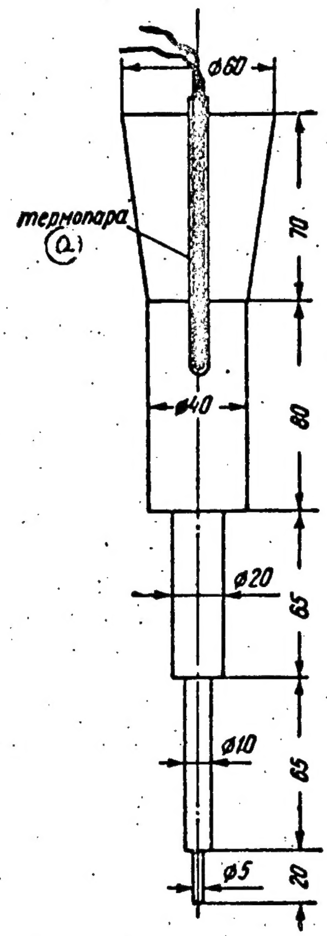

Fig 2. Stepped specimen for the study of the structure with thermocouple inserted for the recording of the rate of cooling.

a - thermocouple

For the determination of the macrostructure, stepped specimens were cast (Fig 2) which were cut along the plane of symmetry and ground. The ground sections were etched with aqua regia (one part nitric acid and three parts hydrochloric acid) or with concentrated nitric acid at 20C. The duration of the etch is given in Table 1. After etching the specimens were carefully washed in running water and then in alcohol.

The grain size was determined by two methods: with the aid of the 16-point reference scale (Table 2) or by means of the computation of the mean section S_{cp} from the equation

$$S_{cp} = F/N,$$

where F is the area under study;

N is the number of grains in the area studied.

In order to exclude the effect of the neighboring sections in the stepped specimen on the macrostructure the grain size was determined in the central portion of each step.

The investigation of the relation of the duration of the solidification of the stepped specimen and of the individually cast cylindrical sample of corresponding dimensions with diameters of 10, 20, and 40 mm indicated that they differed by only 3-5%. Consequently the data obtained characterizes the castings of the corresponding reduced thickness with sufficient accuracy.

The dimensions of the stepped specimen used permitted the variation of the duration of the solidification and consequently of the rate of cooling over a comparatively narrow range. In order to expand this range in the direction of reduction of the rate of cooling and also to approach realistic conditions of casting into ceramic forms using investment casting, the specimens were cast into both cold forms and forms heated to 800C. The cooling in the heated forms reduces the excess temperature θ and is equivalent to an increase of the diameter of the specimens.

Table 1

Duration of Etch, Minutes, of Macrospecimens (Alloys of Fe-Cr-Ni and Ni-Cr Systems)* As a Function of the Content of the Alloying Elements, %

(a) Травители	Al				Co			W			Mo			Nb		Ti	
	1	3	5	10	5	10	15	3	5	10	3	5	10	3	5	1	3
$3HCl + HNO_3$	5	5	5	—	—	—	—	5	5	10	—	5	—	20	25	25	25
HNO_3 концентрированная (b)	—	—	—	10	20	25	25	—	—	—	15	—	5	—	—	—	—

* The duration of the etch of these alloys without alloying elements is 5-20 minutes

a - etchants; b - concentrated

Table 2

Reference Macrostructure Scale

Балл (a)	Площадь зерен, $мм^2$ (b)	Балл (a)	Площадь зерен, $мм^2$ (b)	Балл (a)	Площадь зерен, $мм^2$ (b)	Балл (a)	Площадь зерен, $мм^2$ (b)
1	1,0	9	37,5	5	7,5	13	125
2	1,0	10	50	6	10,0	14	150
3	3,0	11	75	7	17,5	15	225
4	5,0	12	100	8	25,0	16	300 и более (c)

a - points; b - area of grains, mm^2; c - and more

In order to establish the relation between the diameters of the specimens during casting into cold D_x and heated $D_г$ forms we studied their conditions of solidification (duration) and the macrostructure, on the basis of which we established the following relation:

$D_г$, мм . .	10	20	40	60
D_x, мм . .	18	38	70	105

Thus the increase of the form temperature to 800C is equivalent to the increase of the specimen diameter by a factor of 1.75.

The forms for the casting of the specimens were fabricated from the following mixture, %:

Quartz sand 1K025	88
Molding clay	5.6
Liquid glass (sp. wt. = 1.5); M = 2.5	4.6
Fuel oil	0.9
Caustic sulfide	0.9

The forms were dried at 200C for three hours and the forms intended for heating to 800C additionally covered with a paint of the following composition, %

Marshalite	75
Bentonite	3
Dextrin	1.5
Water	20.5

After coating, the forms were dried at 200C for two hours and then heated to 800C.

In the study of the reference alloys the metal was heated in a furnace to 1580-1600C and its temperature during pouring into the form was 1560, 1520, and 1480C. In the study of the actual alloys variations were made in both the pouring temperature and the temperature of the metal in the furnace. Here the superheat above the liquidus amounted to 200C and more.

All the melts were made in an induction furnace using a magnesite lining of the crucible. The following charge materials were used:

Rod steel with 0.12% C
Ferrochrome with 70% Cr, grade Khr 000
Chromium (97-98% Cr)
Granulated nickel
Nickel-carbon alloy with 1.85% C
Aluminum grade AB-000
Molybdenum in slabs
Cobalt and tungsten of high purity
Titanium in briquettes, degassed in a vacuum at 1000C at a vacuum of 10^{-5} mm Hg
Ferro-niobium (58.6% Nb)

The deoxidation of the liquid alloy was accomplished by 0.3-0.4% calcium-silicon grade KaCu-0 before pouring from the furnace.

2. THE PRIMARY CRYSTALLIZATION OF THE HEAT-RESISTANT ALLOYS OF REFERENCE COMPOSITIONS (20% Cr, 0.12 and 0.35% C, 0-80% Ni)

p 11 →

From the state diagrams of the corresponding alloys we know that the alloys containing 20% Ni and more have an austenitic structure for both 0.12% C and 0.35% C and only two alloys with 20% Cr and 0.35% C, containing nickel, have a pearlitic structure.

The data on the size of the grains in the various sections of the stepped specimens cast in the cold and hot forms are shown in Fig 3. They permit us to establish the following relations:

1. The alloys containing no nickel (for both 0.12%C and 0.35%C) have a fine-grained structure. The zone of the columnar crystals is absent in all the sections of the stepped specimens cast in both the hot and cold forms for all the superheats investigated.

2. The introduction of nickel to 60% causes an increase of the grain size. With a Ni content of 80% in the alloy, the grain dimensions are somewhat smaller than with 60% Ni. Regardless of the carbon content and the form temperature, the point corresponding to 60% Ni is the inflection point of the curves characterizing the variation of the grain size with the nickel content in the various sections.

In all the alloys containing nickel in the quantities investigated there is a zone of columnar crystals and this zone is the more strongly developed for the lower cooling rates. This effect of the nickel is explained first of all by the fact that the rate of growth of the crystals of the chrome-nickel austenite significantly exceeds the rate of growth of the crystals of the ferrochrome (Ref 6). In addition, with an increase of the nickel content there is a reduction of the thermal conductivity of the alloy and an increase of the temperature gradient. The presence of a maximum on the curve characterizing the variation of the grain size as a function of the nickel content is explained by the fact that the change of the nickel content has a significant effect on the critical points of the alloy: with an increase of the nickel content to 60% the temperature of the liquidus first lowers and then increases. In this connection, for the conditions used of the melting and pouring the superheat increases along with the increase of the nickel content to 60% and then decreases. This in turn has a considerable effect on the grain size (Fig 4).

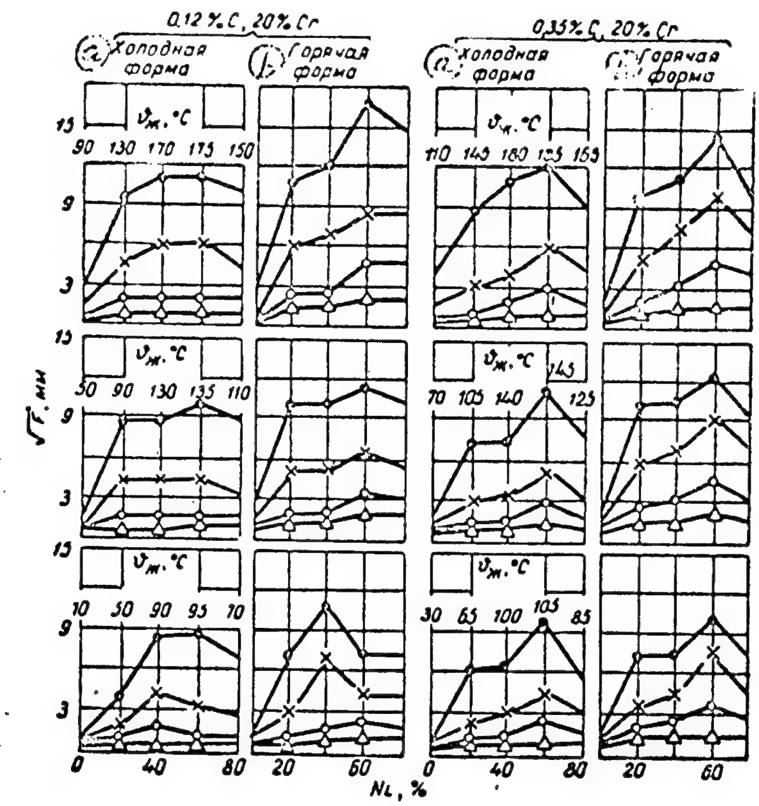

Fig.3. Variation of the grain size of the reference alloys with the composition, the cooling rate, and the superheat of the liquid metal $\vartheta_{ж}$ during pouring (specimen diameters in mm; ● - 40, O - 10, X - 20, Δ - 5)

a - cold form; b - hot form

In relation to the effect of the carbon, we can note that with an increase of the carbon content from 0.12 to 0.35% the grain dimensions in the reference alloys studied vary only slightly. In the sections with diameters of 20 and 40 mm, in the majority of the cases there is observed a tendency toward some increase of the grain size with the indicated variation of the carbon content.

The grain size changes sharply with a change in the section of the stepped specimen. Thus, for example, the grain area in the alloy containing 60% Ni and 0.12% C in the 40 mm section increased during casting into the cold

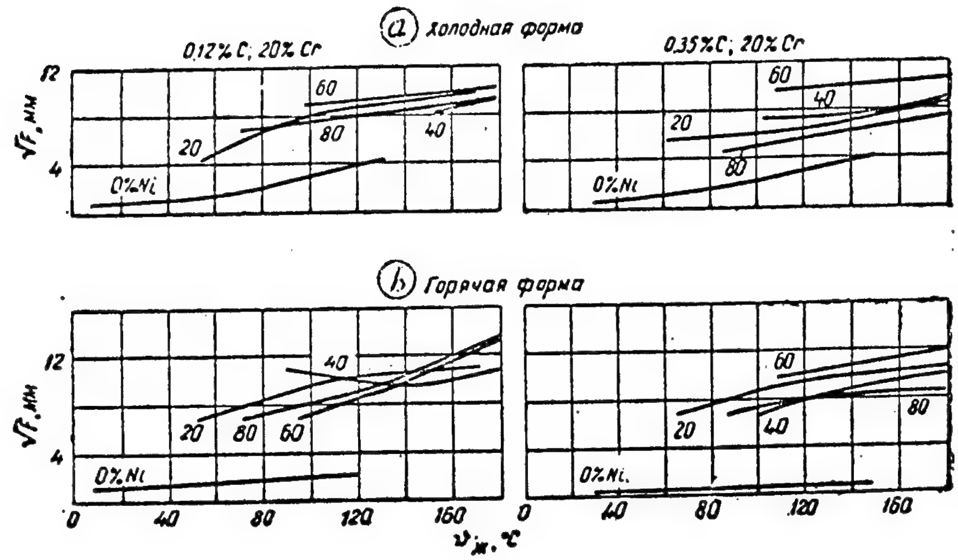

Fig 4. Variation of grain size of the reference alloys with the superheat above the liquidus temperature

a - cold form; b - hot form

form with the maximum pouring temperature from 1 to 125 mm^2 (i.e. by 125 times), and during casting into the hot form from 5 to 300 mm^2 (i.e. by 60 times). In the specimens of the other alloys and also in the specimens poured at other temperatures the relative grain growth was somewhat less (by 25-75 times). This strong effect of the section dimension on the grain size can be explained by the increase in the duration of the solidification.

A similar effect is shown by heating of the form where, as we have mentioned, heating to 800C is equivalent to an increase of the specimen diameter by 1.75 times. However we can note that the effect of the heating of the form is greater for the small diameter specimens. Thus, for example, in the section with diameters of 5 and 10 mm with 60% Ni and 0.12% C in the alloy the area of the individual grains increased by 5 times (from 1 and 5 to 5 and 25 mm^2) while in the section with a diameter of 40 mm the area increased by 2.4 times (from 125 to 300 mm^2).

An increase on the pouring temperature also leads to grain growth (see Fig 3). Thus, for example, in the alloy containing 60% Ni and 0.12% C increasing of the pouring temperature from 1480 to 1560C caused an increase of the grain area in the 40 mm diameter section of nearly 1.8 times (from 75 to 125 mm^2). This effect of the pouring temperature is also explained by the increase of the duration of the solidification of the specimens.

This effect of the dimension of the specimen section, the heating of the form, and the pouring temperature is in good agreement with the variation of the grain size with the duration of the solidification (Fig 5). As we can see from the figure, there is a correlation between these variables independent of the cause leading to the increase of the duration of the solidification. To illustrate the effect of the duration of the solidification, Fig 6 presents the macrostructure of the stepped specimens of the reference alloy with solidification durations of 50 and 170 seconds.

On the basis of the data presented we can conclude that the increase of the nickel content to 60% and the increase of the solidification duration (both as a result of the increase of the specimen dimensions and the heating of the form and as a result of increasing the pouring temperature) leads to a significant increase of the grain size and an expansion of the region of the columnar crystals.

3. THE EFFECT OF ALLOYING ELEMENTS OF THE NICKEL-CHROMIUM REFERENCE ALLOY ON THE PRIMARY CRYSTALLIZATION

As the reference alloy for the study of the effect of alloying elements on the macrostructure we took the alloy with 0.12% C, 80% Ni, 20% Cr. With the introduction of the alloying elements the nickel content was reduced and the chromium content was held constant.

The data obtained are presented in Fig 7. They permit drawing the following conclusions concerning the effect of the alloying elements on the macrostructure of the 0.12/20/80 alloy:

FeB, NiB

1. The introduction of 1% Al somewhat fragmentizes the grain of the reference alloy. With the introduction of 3 and 5% Al the grain size increases while the increase of the aluminum content to 10% fragments the grain to the same size as with 3% Al. FeB, NiB

2. Tungsten to the extent of 3% has no effect on the grain size of the reference alloy while the addition of 5% and especially of 10% of this element leads to sig-

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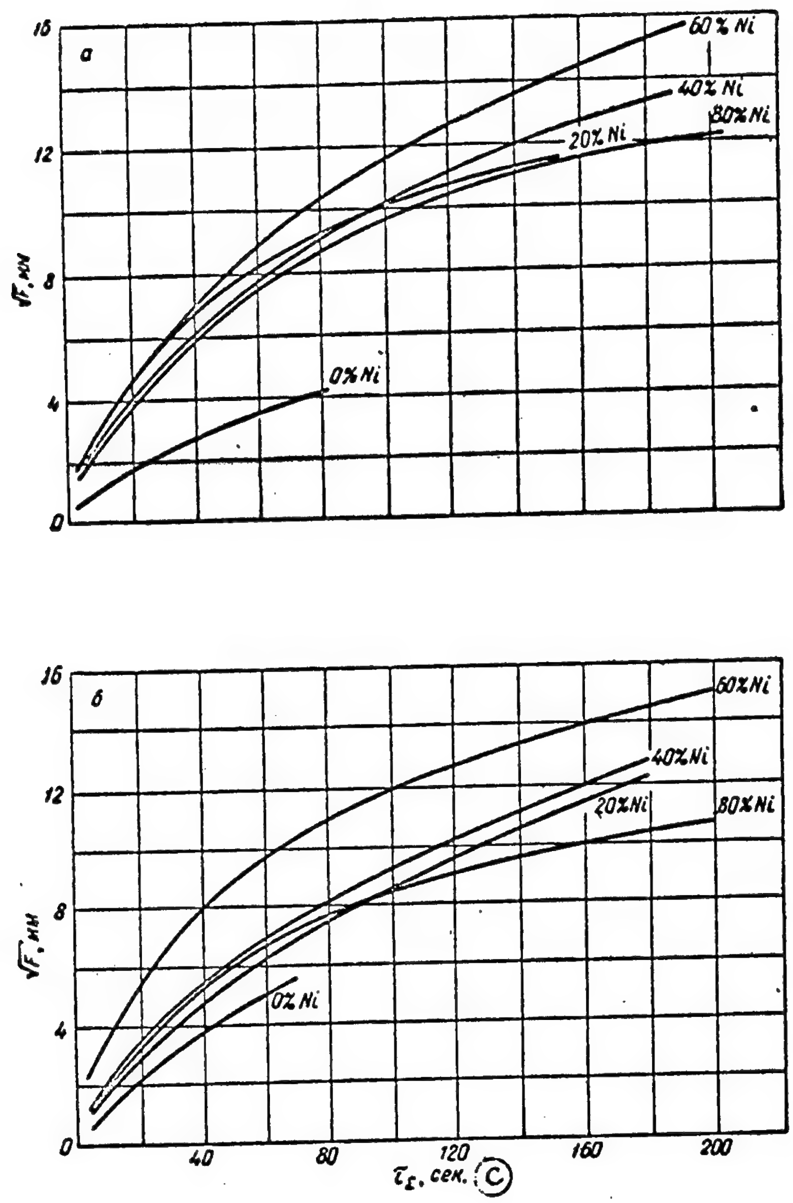

Fig 5. Variation of the grain size in the 40 mm diameter section as a function of the solidification duration τ_{Σ} of the reference alloys:

a - with 0.12% C; b - with 0.35% C; c - sec (Specimens cast into cold and hot forms)

nificant enlargement of the grains. FeB, NiB

3. The introduction of 3% Mo somewhat reduces and 5 and 10% Mo noticeably increases the grain dimensions of the reference alloy. FeB, NiB

4. With 5% Co the grain size of the reference alloy doubles. However, with 10% Co the grain size reduces to the initial value while for 15% Co the grain is half the size of the initial grain.

FeB, Ni Niobium in the amount of 3% leads to an increase of the grain size while 5% of this element has very little effect. FeB, NiB

6. 1% Ti reduces the grain size by a factor of 2.5 times while with 3% Ti the grain size reduces only slightly.

The effect of the elements investigated is not constant but depends on their relative content (Fig 8). With a low content of the alloying elements, for example 1.5-2% (atomic), the strongest effect on the macrostructure of the reference alloy is shown by niobium and titanium with the niobium causing an increase and the titanium a decrease in the grain size. With a medium content of about 4-5% (atomic) the strongest influence is shown by cobalt and molybdenum, resulting in an increase of the grain size. With a high content of the alloying elements of about 10-15% (atomic) the strongest effect is that of aluminum which also increases the grain size. P1

The width of the zone of columnar crystallites in the 40 mm diameter section of the stepped specimen also changes under the influence of the alloying elements (Fig 9). With a content of Al, Mo and Nb of no more than 4% the width of the acicular crystallite zone reduces by 20-30% while in all the other cases it increases.

The columnar crystallite zone is less developed in the smaller diameter sections than in the 40 mm diameter section as a result of the higher cooling rate.

The established effect of the alloying elements on the grain size can also be explained by their influence on the duration of the solidification (Fig 10). There is a direct relation between the change of the solidification duration under the influence of the alloying and the change of the grain size: an increase of the solidification time leads to an increase of the grain size and vice versa. The grain size also depends on the nature of the crystallization process (Fig 11): the larger the crystallization temperature interval, the smaller the ratio of the liquidus time to the totalsolidification time (τ_{π}/τ_{Σ}), the larger the grain will be.

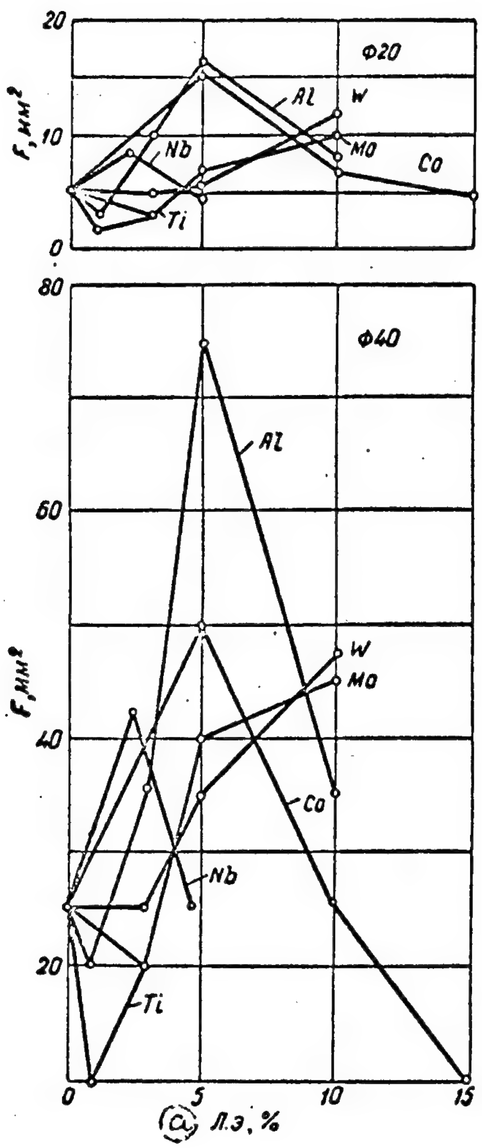

Fig 7. The effect of alloying elements on the grain size in 20 and 40 mm sections of a stepped specimen of the reference alloy 0.12/20/80

a - alloying element

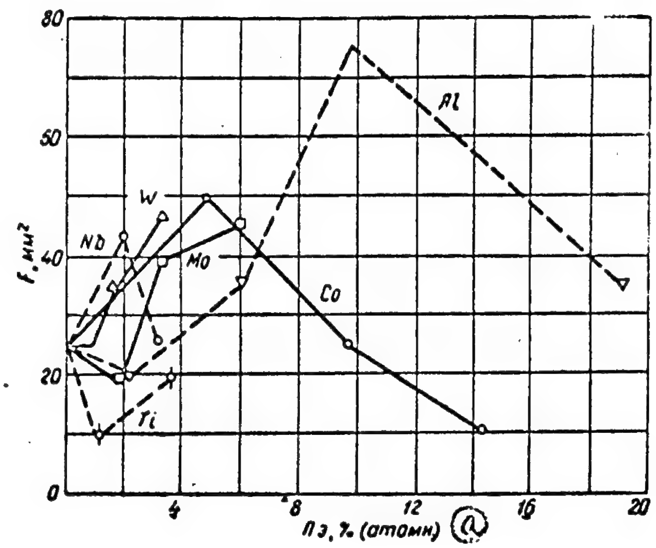

Fig 8. Effect of alloying elements on the grain size in the 40 mm section of a stepped specimen of the reference alloy of composition 0.12/20/80, a - atomic

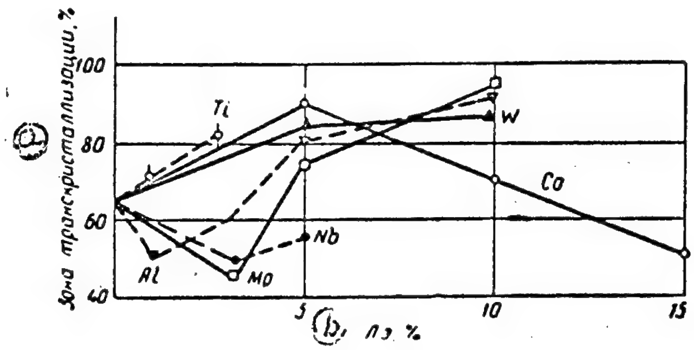

Fig 9. Effect of alloying elements on the width of the transcrystallization zone in the 40 mm section of a stepped specimen of the reference alloy

a - transcrystallization zone; b - alloying element

The width of the transcrystallization zone also is related in a definite manner with the nature of the crystallization process: the longer the crystallization interval and the smaller τ_{Π}/τ_{Σ} the wider the zone of the columnar crystals (Fig 12).

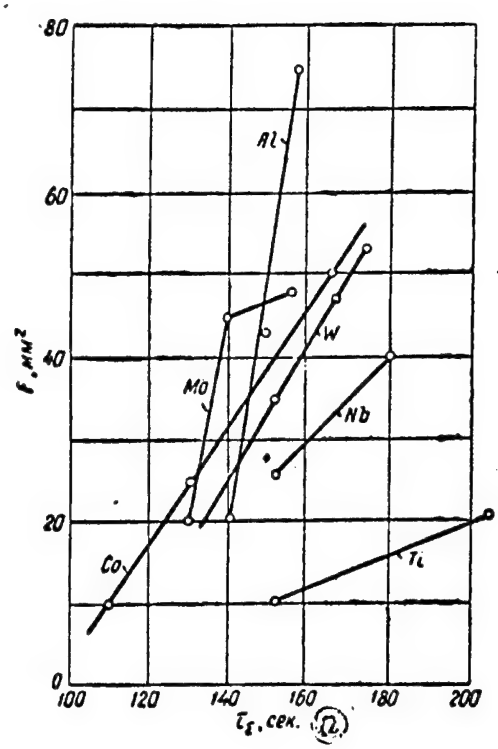

Fig 10. Relation between the grain size in the 40 mm section of a stepped specimen and the duration of the solidification of the reference composition alloy

a - sec

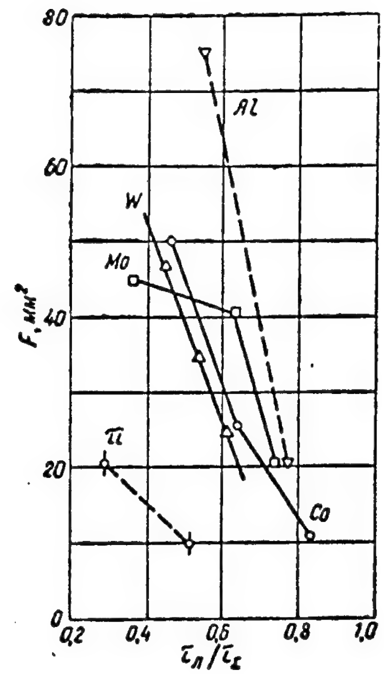

Fig 11. The effect of the relative duration of the liquidus on the grain size in the 40 mm section of the stepped specimen of the reference alloy

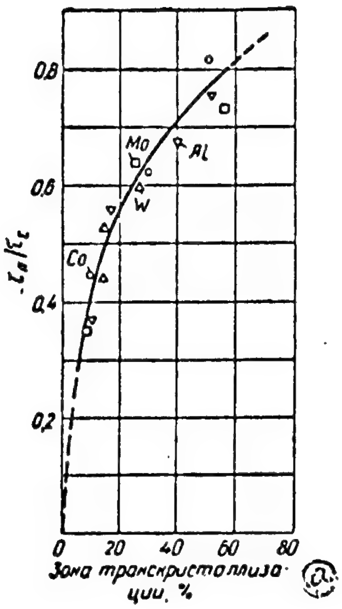

Fig 12. The effect of the relative duration of the liquidus on the width of the transcrystallization zone in the 40 mm section of the stepped specimen of the reference alloy

a - crystallization zone

4. PRIMARY CRYSTALLIZATION OF THE HEAT-RESISTANT ALLOYS OF PRACTICABLE COMPOSITIONS

FeB, N.B

We investigated the primary crystallization of the alloys X1, X32, 111, LA3, EI612, and No 6 and their sensitivity to the variation of the cooling rate and the change of the liquid state. (The alloys X1, X32, 111 (developed in the casting laboratory of Leningrad Poly Inst) and LA3 (developed in the Central Sci. Res. Inst. of Technology and Machine Design and in the Central Boiler and Turbine Institute) are austenitic, on an iron base; the EI612 alloy has an iron-nickel base, while No 6 is a Nimonic type on a nickel base).

N.B

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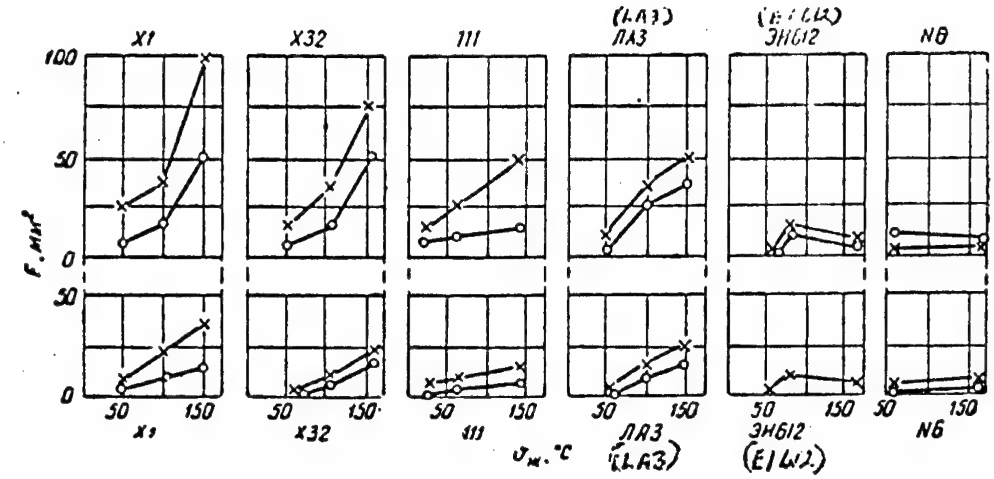

Fig 13. Variation of the grain size of the heat-resistant alloys of practical compositions with superheat $\Theta_{ж}$ and cooling rate in the 40 mm (above) and 20 mm (below) sections; x - hot forms; O - cold forms

Just as in the case of the reference alloys, the rate of cooling was varied by means of the change of the thickness of the specimens, the form temperature, and the pouring temperature of the metal. The macrostructure was studied in the 20 and 40 mm sections (Fig 13). The X1 and X32 alloys had the coarsest grains and were most sensitive to the cooling rate while the No 6 and EI612 alloys were least sensitive. Thus, for example, the increase of the pouring temperature of the X1 and X32 alloys by 150C leads to a 10 fold increase in the grain size in the 40 mm section (from 5 to 50 mm^2) while for the No 6 alloy the grain size remains constant for this change (5 mm^2).

The variation found for the crystallization as a function of the cooling rate is explained, just as in the case of the reference alloys, by the change of the duration of the solidification: an increase of the solidification time leads to growth of the grain. An exception is the EI612 alloy in which the increase of ~~the~~ superheat during pouring from 85 to 175C above the liquidus and the increase in this connection of the solidification period leads to a reduction of the grain from 10 to 5 mm^2. This effect is probably caused by the fact that the superheat to 175C for the EI612

P21

Table 3

Grain Size in Specimens of Heat-Resistant Alloys of Various Types

Сплав (a)	Температура перегрева сплава в печи, °C (b)	Температура перегрева жидкого сплава в ковше, °C (c)	Площадь зерен, мм² (d): (e) холодная форма, ø 40 мм	(e) холодная форма, ø 20 мм	(f) горячая форма, ø 40 мм	(f) горячая форма, ø 20 мм
X1	0	0	5	1	—	—
	200	150	50	15	100	35
		100	15	10	35	25
		50	7	5	25	7
X32	0	0	5	1	—	—
	200	150	50	15	75	25
		105	25	3	35	10
		55	10	1	15	1
№ 6	0	0	5	1	—	—
	200	170	5	3	5	3
		115	5	1	3	3
		70	5	1	3	1
111	0	0	5	3	—	—
	200	140	15	5	50	15
		60	10	3	25	7
		25	7	3	15	5
ЛА3 (LA3)	0	0	5	1	—	—
	200	150	35	15	50	25
		100	25	3	35	15
		50	1	1	10	1
ЭИ612 (EI612)	0	0	15	5	—	—
	200	175	5	10	5	7
		80	10	7	15	10
		60	1	1	1	1

a - alloy; b - superheat temperature of alloy in furnace; c - superheat temperature of liquid alloy in ladle; d - grain area; e - cold form; f - hot form

alloy exceeds the critical value.

It has been established (Table 3) that the increase of the superheat of the metal in the furnace to 200C for all

of the alloys studied except EI612 leads to growth of the grain. A higher superheat leads to a reduction of the grain size not only in the EI612 alloy but also in the [illegible], X32 and LA3 alloys. In the 111 alloy the grain size continues to increase even with superheat of more than 200C while in the No 6 alloy the grain size does not change under the influence of the superheat of the metal in the furnace.

The fractionation of the grain with a high superheat (in spite of the increase of the duration of the solidification) is caused by the alloy solidifying with considerable overcooling. The increase of the supercooling during solidification is caused by the "purification" of the alloy during the high superheat, i.e. the reduction of the residual inclusions which can serve as crystallization centers.

CONCLUSIONS

FeB,NiB

1. The grain size in both the iron-base and the nickel-base heat-resistant alloys varies significantly as a function of the alloy composition, the duration of solidification and the liquid state.

2. The iron-base alloys containing 0.12% and 0.35% C, 20% Cr and containing no nickel have a fine-grained structure and are free of the transcrystallization zone in castings of small section. The addition of nickel to the amount of 60% causes a continuous coarsening of the grain and the formation of a zone of columnar crystals in all sections.

3. The effect of the alloying elements Al, W, Mo, Co, Nb and Ti on the grain size in the reference composition alloy (20% Cr, 80% Ni, 0.12% C) differs and depends on the content of the alloying elements. The strongest effect at a concentration of 1.5-2% (atomic) is shown by niobium and titanium where niobium increases and titanium decreases the grain size of the reference alloy. At a concentration of 4-5% the strongest effects are shown by Co and Mo which increase the grain size while at concentrations of 10-15% the strongest effect is that of aluminum which also increases the grain size.

The zone of the columnar crystals diminishes with the presence in the alloy of no more than 4% Al, Mo or Nb. Under the influence of the other elements and also for large concentrations of Al, Mo and Nb the width of the zone of the columnar crystals increases.

4. The variation of the grain size under the influence of the various elements is explained by their effect on the duration of the solidification and the nature of the crystal-

lization process: the elements which increase the duration of the solidification and the crystallization temperature interval tend to increase the grain size, while the elements which reduce the values of these parameters reduce the grain size.

The width of the columnar crystal zone depends primarily on the nature of the crystallization process: the larger the crystallization temperature range, the wider this zone is.

5. The grain size and the width of the columnar crystal zone in the alloys of reference composition change significantly as a function of the superheat, the pouring temperature of the metal, the form temperature and the section diameter of the specimens. With an increase in the absolute magnitude of these parameters the grain size and the width of the columnar crystal zone increase as a result of the increase in the solidification time.

6. The grain size and the width of the columnar crystal zone of the practical heat-resistant alloys investigated -- X1, X32, 111, LA3, EI612 and No 6 -- also depend on the composition, the melting and pouring conditions, the form temperature and the specimen cross section.

The general relations established for the alloys of the reference composition are also valid for the practical heat-resistant alloys. A peculiarity of certain of these alloys which has been established in this study is the variation of the grain size as a function of the superheat. In the EI612 alloy a super heat of more than 100C leads to fractionation of the grain. For the X1, X32 and LA3 alloys the critical value of the superheat above which the grain fractionates is higher than 200C.

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THE STRUCTURE AND THE PROPERTIES OF THIN-WALL CASTINGS FROM THE HEAT-RESISTANT ALLOYS

/Following is the translation of an article by M.T. Bogdanov in the Russian-language publication Trudy LPI (Transactions of Leningrad Poly. Inst.), No 224, Moscow, 1963, pp 177-194./

The present paper covers an investigation of the forming of cast gas turbine blades from heat-resistant alloys as a function of such technological casting conditions as the temperature of the metal, the position of the casting in the form, the form material and form heating, the method of delivery of the metal. A concurrent study was made of the effect of the structure, density, gas content and scaling on the mechanical properties of the casting at normal and elevated temperatures.

The experimental castings were poured using the heat-resistant austenitic steel X1 developed at LPI (type EI572) and the nickel-base heat-resistant alloy No 6 developed for gas turbine blades. In order to evaluate the mechanical properties we produced both very thin-walled castings in the form of test specimens 6 mm in diameter (tensile test) and with 11x11 mm section (impact strength) and also the heavier cloverleaf-like castings (clovers) with lobes 14 mm thick from which the corresponding specimens were cut. The dimensions of the test samples used correspond approximately to the dimensions of the working section and the shank section of medium-size turbine blades (Ref 1).

The forms for the production of the thin wall castings were filled from above (Св), by siphon (Сф) using primarily either single-side (Бс) or dual-side (Бг and Бв) supply of the metal. The ceramic forms for the precision cast thin-wall specimens had temperatures of 800, 400, and 20C and those for the cloverleafs had temperatures of 800 and 20C.

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In addition, the cloverleafs were cast into cast iron forms to provide for high cooling rates.

The X1 steel was poured with superheat above the liquidus of 250, 160 and 75C while 300, 200 and 100C superheats were used for the alloy No 6. For the thin-wall castings account was taken of the cooling of the metal during the flow in the feeding system. The forms were fabricated by the lost wax method. To increase the strength of the ceramic covering of the forms and to eliminate cracks in this coating during cooling to 400 and 20C, 3% boric acid was added to the filler and slow cooling of the forms together with the roasting furnace was used. The melts were made in induction furnaces with capacities of 12 and 25 kg with magnesite lining. The X1 steel was melted from the basic charge materials, the alloy No 6 from commercial billets. The density of the metal was determined by hydrostatic weighing of the specimens in air and in toluene. The macrostructure of the X1 steel was exposed by etching the sections in aqua regia while that of the alloy No 6 was etched in the following reagent: 150 g of $CuSO_4$, 500 cm^3 of concentrated HCl, 35 cm^3 of concentrated H_2SO_4. The microstructure of the X1 steel was exposed by electrolytic etching in a 10% solution of oxalic acid and that of alloy No 6 by Marble's reagent.

The design methods (Refs 2,3) which permit analytic evaluation of the effect of metal superheat, form heating, metal supply, position of the casting in the form (for given thermal and physical properties), dimensions and shape of the castings and forms on the thermal conditions of the casting process indicated (Ref 4) that the overall differential in the rates of cooling of the thin-wall and the cloverleaf castings reached a considerable magntidue in the test conducted. For the thin-wall castings the rate of cooling varied by a factor of about 7 and for the cloverleaf about 12. This provided a significant difference in the structure and the density of the X1 steel and the No 6 alloy which was sufficient for the analysis and the comparative evaluation of their influence on the mechanical and the heat-resistant properties of the castings.

The grain size of the X1 steel, which has a natural coarse-grained structure, was varied by more than 600 times (in the thin-wall castings from 0.2 to 19.6 and in the cloverleafs from 3 to 128 mm^2); the density varied by 0.1402 g cm^3 (in the thin-wall castings from 7.7472 to 7.8411 and in the cloverleaf from 7.8331 to 7.8874 g/cm^3). The grain size and the density of alloy No 6 were varied by about 30 times and by 0.2422 g/cm^3 (in the thin-wall

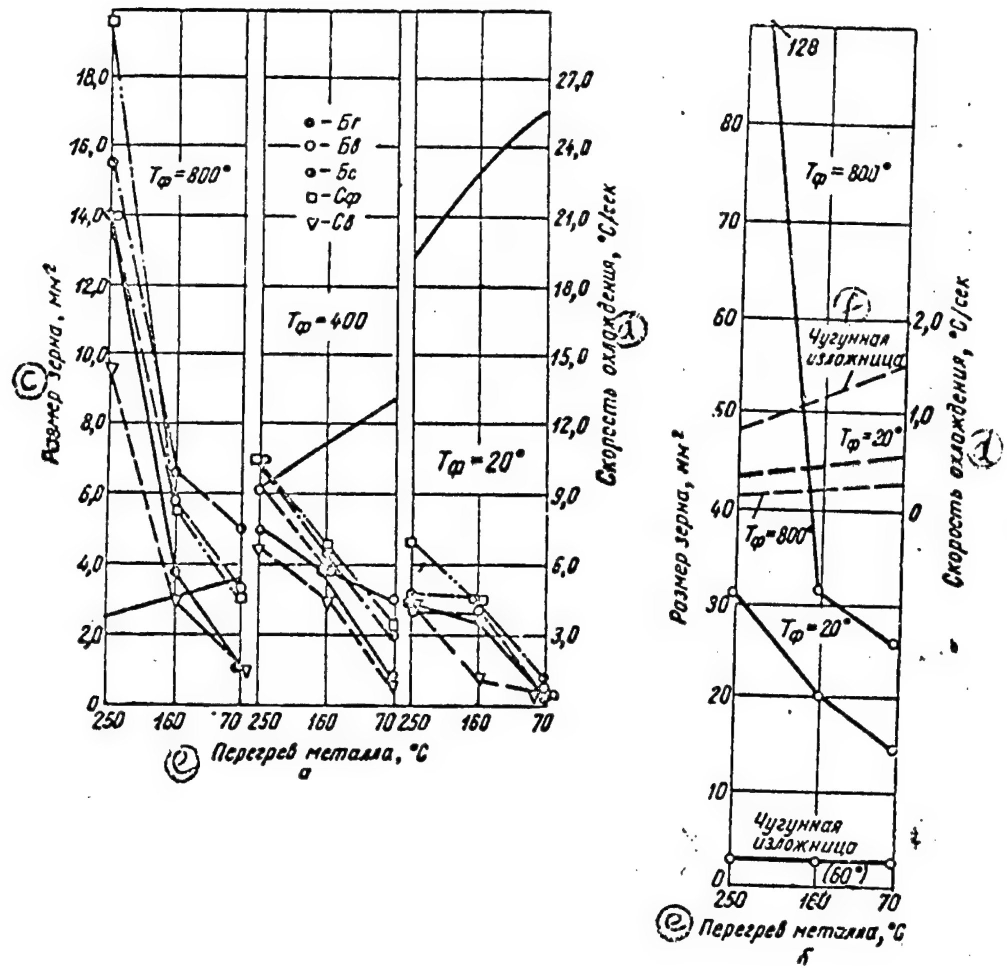

Fig 1. Grain size of the X1 steel in the thin-wall (a) and cloverleaf (b) castings

c - grain size; d - cooling rate, °C/sec; e - metal superheat; f - cast iron mold; subscript ф - form

castings from 0.2 to 3.5 mm^2 and from 8.0382 to 8.1481 g/cm^3; in the cloverleafs from 1 to 6.5 mm^2 and from 8.2324 to 8.2804 g/cm^3). The hydrogen content in the castings varied by about a factor of 2.

During the casting into the hot molds the reduction of the metal superheat from a high to a medium value causes an especially noticeable change of the structure of the X1

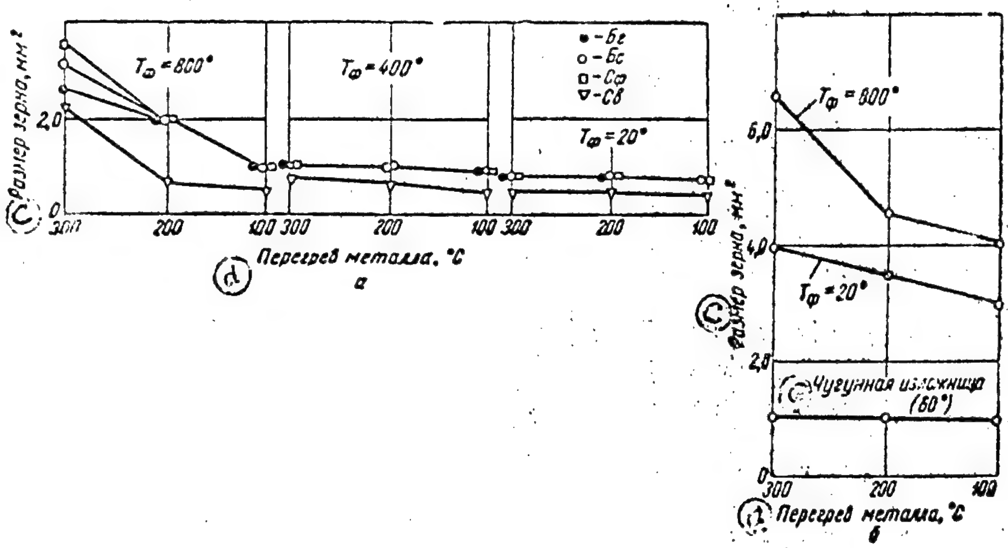

Fig 2. Grain size of the No 6 alloy in thin-wall (a) and cloverleaf (b) castings
c - grain size; d - metal superheat; e - cast iron mold

steel and the No 6 alloy in both the thin-wall and the heavier cloverleaf castings (Figs 1 and 2). In the thin-wall X1 steel castings there is observed a reduction of the grain size by more than a factor of 3 (for bilateral metal feed, from 14 to 3.8 mm^2) while in the cloverleaf castings this reduction is even more pronounced (from 128 to 32 mm^2). Further reduction of the superheat to 75° reduces the grain size of the X1 steel to a lesser degree. In cold forms and forms heated to 400C this severe reduction of the grain size of the X1 steel and of the No 6 alloy with a reduction of the superheat was not observed. The metal superheat does not have any significant effect on the grain size of the alloys in the cloverleaf castings produced in the cast iron mold.

For the X1 steel which is sensitive to the cooling conditions in the ceramic form there is a characteristic change not only of the size but also of the shape of the grain as well as the presence in the section of the cloverleaf castings of zones with differing structure of the metal (Fig 3). The cloverleafs cast in the hot form (800C) with high metal superheat (250C) have a coarse, randomly oriented structure, with medium superheat (160C) the structure is columnar, and with minimum superheat (75C) they have a fine-grained, randomly oriented structure. With a reduction of

the superheat when pouring into cold forms the columnar structure gradually transforms to the two-zone type; in the central portion of the castings there appear smaller equiaxed grains. In the cast iron molds at all the metal superheats there are observed fine acicular crystallites. The cloverleafs cast from the No 6 alloy in all cases have only randomly oriented equiaxed grains while the cloverleafs cast in the cast iron molds have in addition a very small surface zone of thin fine columnar crystallites.

With a increase of the rate of cooling of the thin-wall castings there is a noticeable reduction of the density of the X1 steel regardless of the casting method (Fig 4a) as a result of the reduction of the form heat (from 800 to 20C). However in the range of the reduction of the form temperature from 800 to 400C and the superheat from 250 to 160C there is no significant change observed in the density of the X1 steel. This result of the investigation gives a definite indication of the possibility of the production of high-quality thin-wall castings (to 5 mm) with high density from the austenitic steel (type X1) only under conditions of pouring into hot forms (at least 400C) and with a sufficiently high metal superheat (150-250C).

However for the nickel-base alloys (type No 6) an even higher superheat is required (200-300C) with a consistently high density during casting into a cold form being achieved only with a superheat of 300C (Fig 4b). Pouring of the metal with a superheat of less than 200C inevitably leads to a considerable reduction of the density of the thin-wall castings. It is only with high form temperature (800C) that this reduction of density is less marked.

The density variation of the metal in the more massive castings of the cloverleaf type (Fig 5) has a different nature. With an increase in the cooling rate due to a reduction of the form temperature (from 800 to 20C) or a reduction of the superheat (no less than the average when using a cold form) the density of the X1 steel and the No 6 alloy increases. In the cast iron forms there is noted a tendency toward some reduction of the density of the X1 steel and the No 6 alloy with any reduction of the superheat.

Analysis of the experimental data obtained indicates (Fig 6) that the positive effect of the increase of the rate of cooling is only noted up to a definite critical limit above which the density of the X1 steel (and also of the No 6 alloy) begins to decrease.

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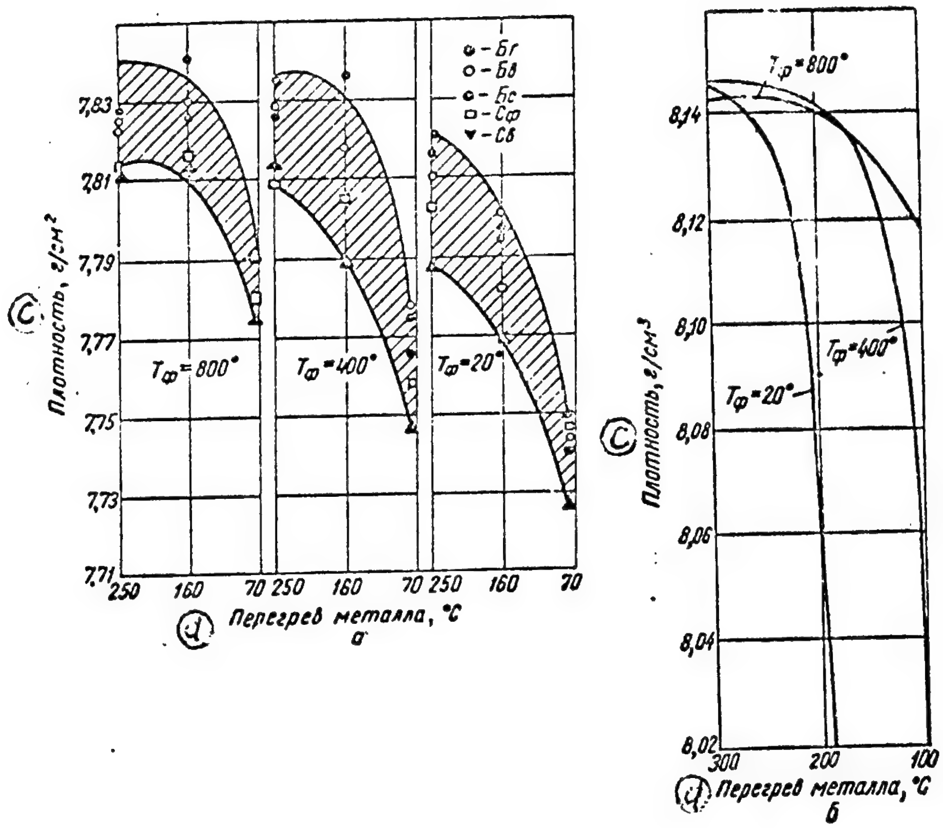

Fig 4. Density of X1 steel (a) and No 6 alloy (b) in thin-wall castings

c - density, g/cm3; d - metal superheat

The established variation of the structure and the density of the thin-wall and the cloverleaf castings as a function of the variation of the superheat cannot be explained solely by the difference of the thermal conditions during casting or by the variation of solely the cooling (solidification) rate. Calculations show (Fig 4) that a significant variation of the superheat does not cause an equally significant change in the cooling rate and the solidification time. With a reduction of the superheat of the X1 steel from 250 to 75C, i.e. by a factor of three, the rate of cooling of the thin-wall castings produced in the hot form increases from 3.8 to 5.5°C/sec, i.e. by about a factor of 1.5. However, with a reduction of the form

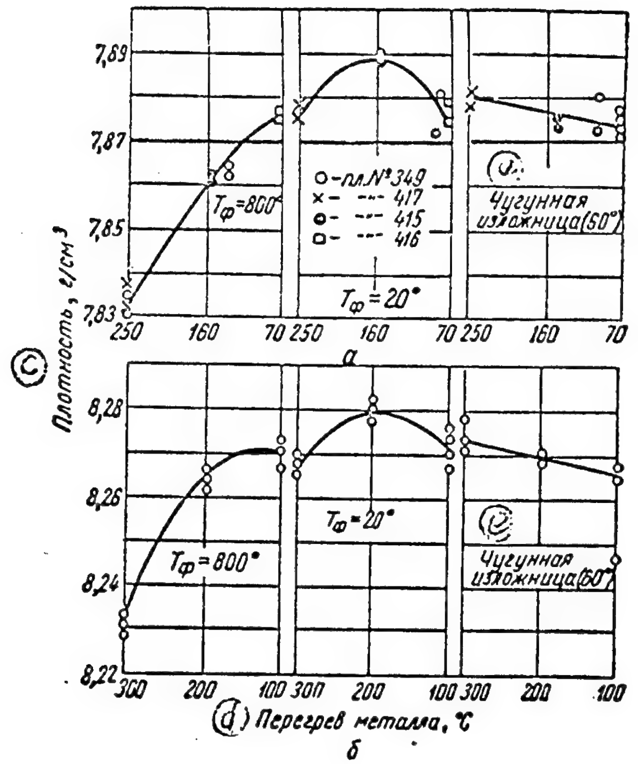

Fig 5. Density of the X1 steel (a) and the No 6 alloy (b) in the cloverleaf castings

c - density, g/cm^3; d - metal superheat; e - cast iron mold;

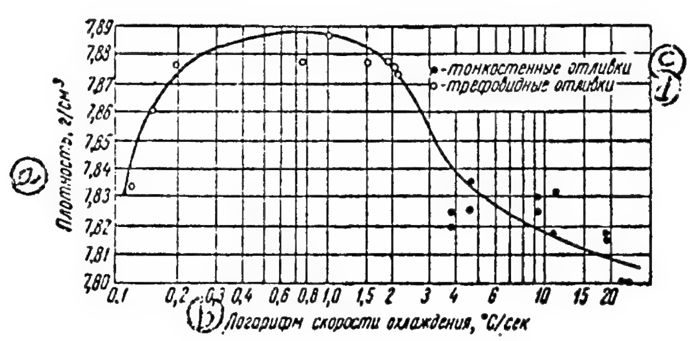

Fig 6. Variation of the steel density as a function of the cooling rate

a - density, g/cm^3; b - logarithm of the cooling rate, $^{\circ}$C/sec; c - thin-wall castings; d - cloverleaf castings

heat from 800 to 400C the cooling rate increases from 3.8 to 9.3°C/sec or by a factor of 2.5. With a further reduction of the form temperature to 20C the cooling rate now increases by about a factor of 5. Thus, considering the comparatively limited possibility for the increasing of the metal temperature in comparison with the increase of the form temperature the more effective method for the changing of the cooling rate of such castings is the heating of the form, although its effect is less powerful than an equivalent absolute magnitude of the metal superheat. At the same time the decisive effect of the superheat is emphasized by the very noticeable change of the structure and density of the castings, not achieved even by a large variation of the cooling rate by variation of form heating.

Superheating, which determines the physical and chemical properties of the liquid metal, exercises a more significant effect on the formation of the structure and the properties of the casting than would be expected solely on the basis of the considerations of the evaluation of the results achieved in the change of the physical heat content of the metal. With an increase of the superheat there are changes in the viscosity and the structure of the liquid metal, the gas content, the amount of nonmetallic inclusions and scale; the suspended nonmetallic inclusions are broken up, partially dissolved or floated out. The metal becomes, according to the definition of A.A. Baykov, physically more "transparent". There is a noticeable deactivation of the nuclei which aids in the enlargement of the primary grain. The X1 steel is deactivated considerably easier than the No 6 alloy which contains a significant quantity of Al and Ti which, by contaminating the liquid metal with a "cloud" of dispersed refractory inclusions (oxides, carbides and nitrides of titanium and aluminum), aid in the conservation of the fine-grain structure even with a high superheat.

The effect of the deactivation, particularly noted in castings produced in the hot molds, decreases with a reduction of the form heating as a result of the more intense cooling of the metal on contact with the walls of the form even during the pouring process.

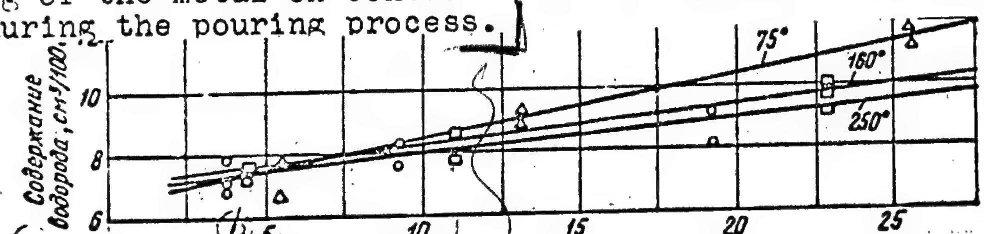

Fig 7. Hydrogen content in castings from X1 steel
a - hydrogen content; b - cooling rate, °C/sec

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The variation of the metal superheat can, it seems, also affect the content and the conditions for the evolution of the gases from the casting. With a reduction of the temperature of the liquid metal the solubility of the gases, particularly the hydrogen, is reduced. Therefore we would expect that with a reduction of the metal superheat the hydrogen content in the castings would also be reduced. However, analyses show that, on the other hand, it increases and even increases more noticeably the higher the rate of cooling of the thin-wall castings (Fig 7) (the hydrogen content was determined for vacuum heating using the method developed in the casting laboratory).

The hydrogen content in the thin-wall, rapidly-cooled castings from the X1 steel reaches 11-12 $cm^3/100$ g, i.e. it is quite high not only in absolute value but also relatively speaking (almost twice as high as in the more slowly cooled castings from the same steel). With this or even a higher hydrogen content the process of the bubbling evolution of the hydrogen from the metal will obviously have a limited development. But the diffusive process for the evolution of the hydrogen can also aid in the formation of interdendritic, intercrystalline openings and pores. We know that the diffusing atomic hydrogen, associating into molecules on contact in the shrinkage pores or other discontinuities, increases the pressure at these locations which then leads to a corresponding increase of the volume of the pores and in all cases hinders the penetration of the liquid metal into these pores during feeding. The increased hydrogen concentration is one of the primary causes for the degradation of the plastic properties of the metal and the formation of various flaws in the castings. Under conditions of a high cooling rate the reduction of the superheat, which sharply reduces the fluidity, degrades the conditions for the filling of the form. This also aids in the formation of unfilled pores in the castings.

The data of Ref 4 show that in the production of thin-wall castings in cold forms and forms heated to 400C comparatively slight superheats (to 100C) can be entirely lost during the flow of the metal in the system for the filling of the form. The filling of the form with minimum superheat leads to the situation where the calculated temperature of the metal in the section of the casting in question is lower than the liquidus temperature (1425C). However, the temperature of the moving stream cannot fall below the temperature of initiation of crystallization in connection with the fact that the solidification process

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encountered during the cooling of the metal aids in maintaining the temperature at a constant level (Ref 2). Therefore we can assume that in the two cases mentioned (temperature of the form of 400 and 20C) the superheat is nearly entirely removed during the flow of the metal in the system for filling of the form, even more rapidly in the case of pouring into a cold form.

We should note that the gas and air bubbles, the nonmetallic inclusions, and the scale which occur during the pouring process and which cannot be floated out because of the high viscosity of the metal (low superheat) or the high cooling rate can also have an unfavorable effect, reducing the density and degrading the mechanical properties of the castings.

The reduction of the density with an increase of the metal viscosity and the cooling rate of the thin-wall castings, which is observed more clearly in the No 6 alloy than in the X1 steel, is explained by the formation of coarser scale in the No 6 alloy. Therefore the use of high superheats of the metal and high temperatures of the forms improves the conditions for the production of sound castings from this alloy.

In the fractures of the thin-wall castings of both of the alloys, poured at high metal superheats into hot forms, there are most frequently encountered individual small blisters of rounded shape. With a reduction of the temperature of the form and particularly of the superheat the rounded form of the blisters disappears and the dimensions increase.

In the X1 steel containing up to 0.2% Ti the blisters (dimensions of no more than 0.5 mm) are detected only in rare cases, primarily with low superheats. In the No 6 alloy the blisters are located in the form of separated segments or even a continuous band along the surface of the castings. The color of the blisters varies: in the X1 steel from golden to light or dark gray, in the No 6 alloy from light-golden to orange and gray, at places with a greenish tint, which is apparently explained by the differing thickness of the blisters.

The quality of the castings made from the scale-forming alloys by free pouring in air also depends to a considerable extent on the method of supply of the metal and the position of the casting in the form. The presence of counter-currents of the metal (during combined two-sided feed and a horizontally positioned casting) or counter currents of the metal and the air (pouring from above) lead to the degradation of the thin-wall castings by scale. For

the No 6 alloy the best results were obtained when using primarily single-sided lateral feed or siphon feed of the metal.

Thus the density of the thin-wall castings depends on several complex processes taking place during the pouring and the cooling. During the flow of the metal in the mold-supply system, along with the cooling of the metal and the evolution of gases there also occur the reverse processes of the absorption of gases, the oxidation of the metal, the mechanical entrainment of air, gases and nonmetallic inclusions.

The results obtained from the investigation make it possible to analyze the effect of the structure, density and scale on the formation of the mechanical properties in the thin and more massive portions of the castings from the X1 steel and the No 6 alloy. P40

Fig 8 presents the variation of the plastic properties (δ,%) for the X1 steel as a function of the density of the thin-wall castings. Analysis of the data obtained indicates that the difference in the structure has a comparatively limited effect on the plasticity (even when using the X1 steel which is relatively sensitivity to rate of cooling). However the effect of the density is a decisive factor in all cases. Between the density and the plastic properties there is established a direct relationship, somewhat distorted for the X1 steel with by the reduced values of the plastic properties of the castings produced with a high metal superheat into a hot form. This fact is explained by the very considerable variation of the microstructure of the X1 steel. The hardening phases, primarily carbonitrides, which are distributed along the boundaries of the grains and within the grains become coarser with a reduction of the cooling rate. Even the appearance of eutectic formation along the grain boundaries is possible. However, in this region of reduced values of the relative elongation there is still observed a definite relationship between the density and the plastic properties.

For alloy No 6 which consists of a solid solution and a complex intermetallic hardening phase, there is not observed a very noticeable change of the microstructure with a change in the pouring conditions. The cases noted of the distortion of the correspondence between the density and the plastic properties are due primarily to the presence in the castings of blister of the oxides and nitrides of titanium and aluminum. These blisters can be the cause of a reduction of the density of the castings not only as a result

of their low specific weights, but since they are surfaces of separation within the liquid metal the blisters aid in the formation of gas bubbles which are not able to separate out in the case of rapid cooling and high viscosity of the metal. They create flaws near the blisters in the thin sections of the castings in the form of fine pits and local gas porosity.

The structure of the thin-wall castings, varying as a function of the method of supply of the metal, in comparison with the density has practically no noticeable effect on the mechanical properties of the castings. The grain sizes of the X1 steel in the castings produced by pouring from above and by lateral two-sided supply of the metal are approximately identical. The mechanical properties (lower for the pouring from above) are determined only by the different density of the castings. The decisive influence of the metal density is shown by the fact that the methods of pouring which provide the highest density simultaneously give the highest mechanical properties of the thin-wall castings at normal and at high temperatures in spite of their differing structure (Refs 1,6). The study of the fractures of the specimens indicates that most frequently the fracture failure of the castings produced by the siphon method of pouring occur in the upper, less dense zone having a fine-grain structure. The plastic properties of the metal in thin plates, approximating in dimensions and shape the medium-size turbine blades, are highest in the portions of the castings having the highest density in spite of the quite different structure (Ref 7).

For the alloys which do not form blisters there is a definite advantage to the lateral two-sided supply of the metal, particularly when the castings are placed in the horizontal plane. The use of lateral, primarily one-sided supply or siphon supply provides intermediate properties and the use of pouring from above provides the least satisfactory results. It must be noted, however, that with low superheats and low heating of the forms the lateral two-sided supply of the metal loses its advantage.

The variation of the plastic properties of the X1 steel in the cloverleaf castings as a function of the pouring condition is shown in Fig 9. In the analysis of the results obtained it is seen that the role of the structure of the cloverleaf castings, whose cooling rate is 20-30 times lower than that of the thin-wall castings, increases significantly. Therefore, for example, an identical increase

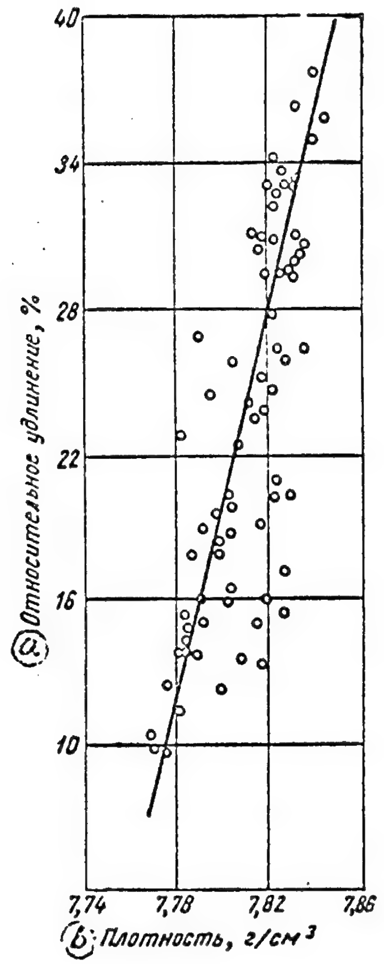

Fig 8. Relation between the density and the plastic properties (δ, %) of the X1 steel in thin-wall castings
a - relative elongation; b - density, g/cm^3

of the density of the X1 steel in the cloverleaf and the thin-wall castings corresponds to a different increase of the plastic properties (lower in the cloverleaf castings which have a less favorable structure, Ref 5). The difference in the structure is explained by the fact that under definite conditions of pouring, the plastic properties of the cloverleaf and the thin-wall castings are practically equivalent, although the density of the cloverleaf castings is significantly higher.

Because of the remaining unsatisfactory coarse-grained structure of the X1 steel in the cloverleaf cast-

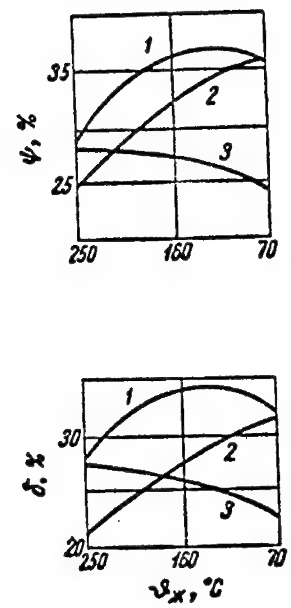

Fig 9. Plastic properties (δ and ψ, %) of the X1 steel in cloverleaf castings

1 - t_{form} = 20C; 2 - t_{form} = 800C; 3 - cast iron form

ings produced in the ceramic forms, the positive effect of the density on the plastic properties is reduced; the correspondence between the variation of the density and the plastic properties is destroyed. This explains why the considerable increase of the density of the cloverleaf castings with a reduction of the heating of the form from 800 to 20C does not cause a correspondingly significant improvement in the plastic properties.

The fine acicular structure of the X1 steel in the cloverleaf castings produced in the cast iron forms is even less favorable than the columnar structure obtained in the ceramic forms. Therefore for similar values of the density the castings produced in the ceramic forms have somewhat higher plastic properties than the castings produced in the cast iron forms.

With the high rates of cooling of the castings achieved in the cast iron mold forms, the level and the uniformity of the plastic properties of the X1 steel depend not only on the density and the structure but also on the unfavorable effect of the flaws which increase with a reduction of the

superheat -- nonmetallic inclusions, gaseous and air bubbles which are mechanically entrained and not evolved from the metal because of the increased viscosity of the melt.

In the cloverleaf castings from the No 6 alloy produced under similar conditions, the effect of the structure is less pronounced. For low rates of cooling, although there is a somewhat less uniform distribution of the precipitation phases within and along the boundaries of the grains, the structure of the metal continues to be quite satisfactory. Therefore the plastic properties of the cloverleaf castings from the No 6 alloy are more dependent on the density. However, with an increase of the rate of cooling and the viscosity of the metal there is an increase in the harmful effect of the blisters. The blisters, which have a decisive influence on the mechanical properties, can significantly reduce the level and the uniformity of these properties.

In the analysis of the results of the mechanical tests there is seen a tendency to some improvement of the strength properties of the X1 steel and the No 6 alloy in the thin-wall and in the cloverleaf castings with an increase in the rate of cooling. For the X1 steel this is particularly noticeable when pouring into a hot form. For the No 6 alloy, as a result of the harmful effect of the blisters with an increase in the rate of cooling and the viscosity of the metal, this tendency is weakened and the strength properties are not uniform.

From Fig 10 we see that the variation of the impact strength of the X1 steel in the thin-wall castings has a somewhat different nature than the variation of the plastic properties; it is less dependent on the density and more dependent on the structure. In the castings produced in the hot forms with a low metal superheat, in connection with the improvement of the structure there is observed FeB, NiB a tendency toward some improvement of the impact strength values. In the forms heated to 400C there is no significant change in the impact strength noted with the exception of a certain reduction as the superheat is reduced to the minimum value. In the cold forms, for any reduction of the superheat below 250C the impact strength decreases as a result of the already significant reduction of the density of the castings. In view of some nonuniformity of the melts, this nature of the variation of the impact strength is more noticeable in the comparison of results of tests of the same melt.

The reduced dependence of the impact strength on the density and also the comparatively slight variation of the structure of the X1 steel due to the method of supply of the

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metal leads to the fact that the different methods of pouring investigated give practically equivalent results with respect to the impact strength. It is only when using siphon pouring with the simultaneous effect of the less favorable structure and the reduced density of the castings that the impact strength is somewhat reduced.

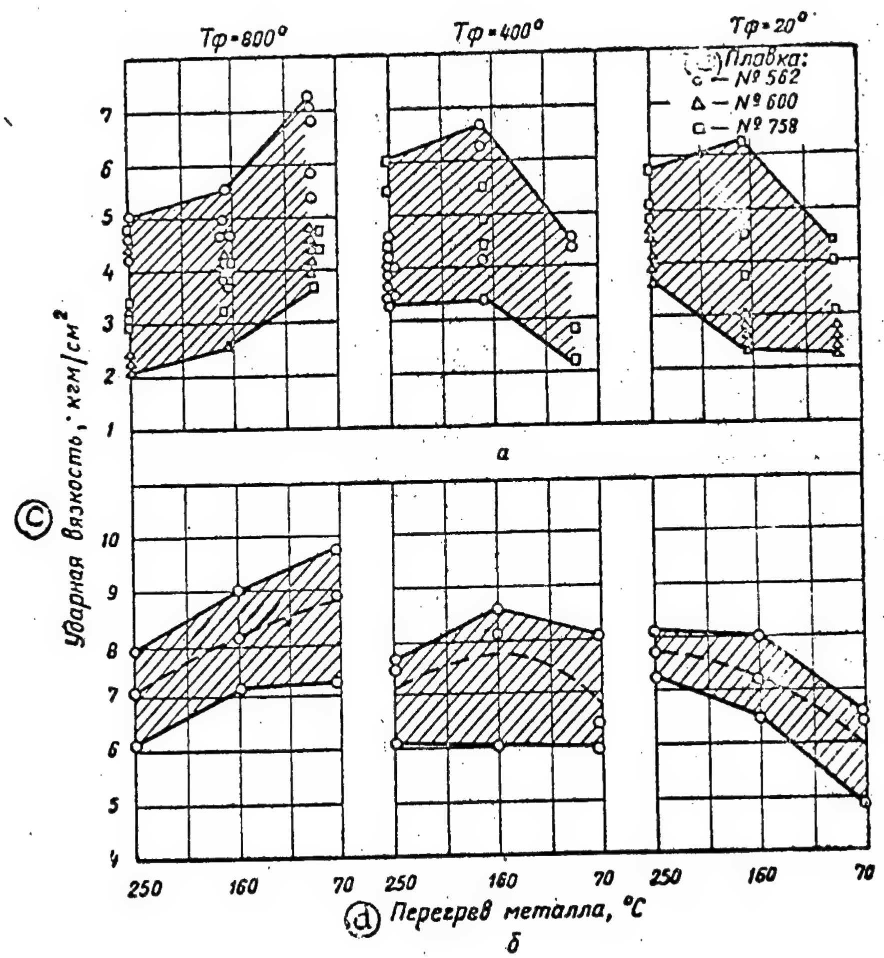

Fig 10. Impact strength of X1 steel in thin-wall castings

a - at 20C; b - at 650C; c - impact strength kgm/cm^2; d - metal superheat; e - specimen No

The variation of the impact strength of the No 6 alloy in the thin-wall castings is found to be in better correspondence with the variation of the plastic properties since the level of both these properties is determined primarily by the effect of the blisters (Fig 11). The variation of the impact strength with the blisters is also characterized by the results of the tests of specimens cast by different methods. The most satisfactory properties were obtained using the siphon method and the lateral, primarily one-sided, method of supply of the metal. The impact strength depends, in addition, on the quantity and the nature of the distribution of the brittle carbonitrides Ti(CN) which are much more poorly separated out, just as the blisters, with an increase of the metal viscosity and increased cooling rate.

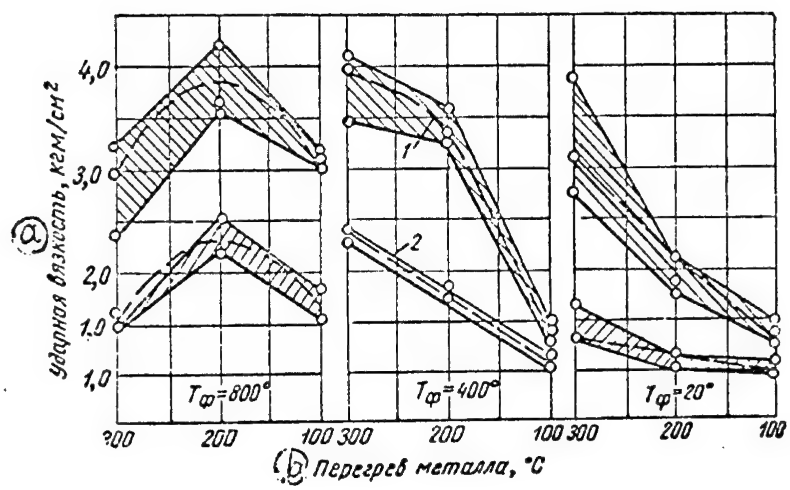

Fig 11. Impact strength of No 6 alloy in thin-wall castings at normal and elevated temperatures:

1 - a_k at high temperature; 2 - a_k at 20C; 3 - impact strength, kgm/cm^2; 4 - metal superheat; a - impact strength; b - metal superheat

In the more massive cloverleaf castings from the X1 steel produced in a hot form, the observed tendency with a reduction of the metal superheat toward an increase of the impact strength at 20C is explained by the simultaneous favorable effect of the improvement of the structure and the increase of the density of the metal. The slight variation of the impact strength of the X1 steel poured into the ceramic form is explained by the favorable effect of the

retained coarse-grain structure of the cloverleaf castings.

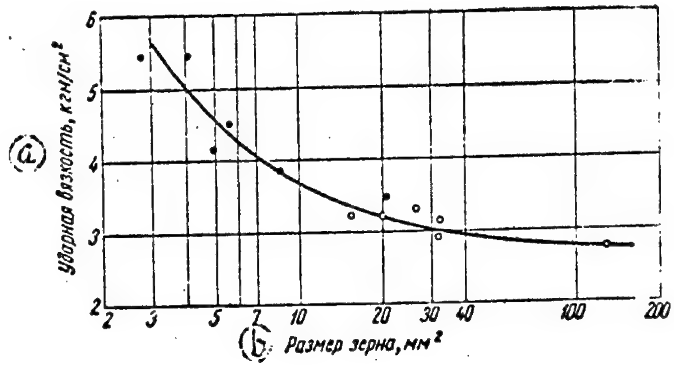

Fig 12. The effect of the grain size on the impact strength of X1 steel

⊙ - thin-wall castings; O - cloverleaf castings; a - impact strength, kgm/cm^2; b - grain size

On reaching some critical grain size the effect of the structure on the impact strength is seen to be more intense and this is also observed in the thin-wall castings (Fig 12). The appearance in the thin-wall castings of columnar crystallites directed along the thermal flow (pouring into a cold form with high and medium superheats), and in the cloverleaf castings (pouring into the cast iron mold forms) of even finer acicular crystallites reduces the impact strength. The fracture of the specimens takes place along the linear boundaries of the crystallites weakened by the brittle precipitates which are not destroyed by heat treatment.

The variation of the impact strength of the No 6 alloy in the cloverleaf castings with a reduction of the form heating and of the superheat depends not only on the moderate change of the structure and the density but also on the increasingly harmful effect of the blisters with an increase in the cooling rate and the metal viscosity. Therefore the tendency toward the increase of the impact strength which occurs in connection with the simultaneous improvement (simultaneously with the reduction of the grain size there is a reduction of the size of the titanium carbide inclusions, Ref 9) of the structure and the increase

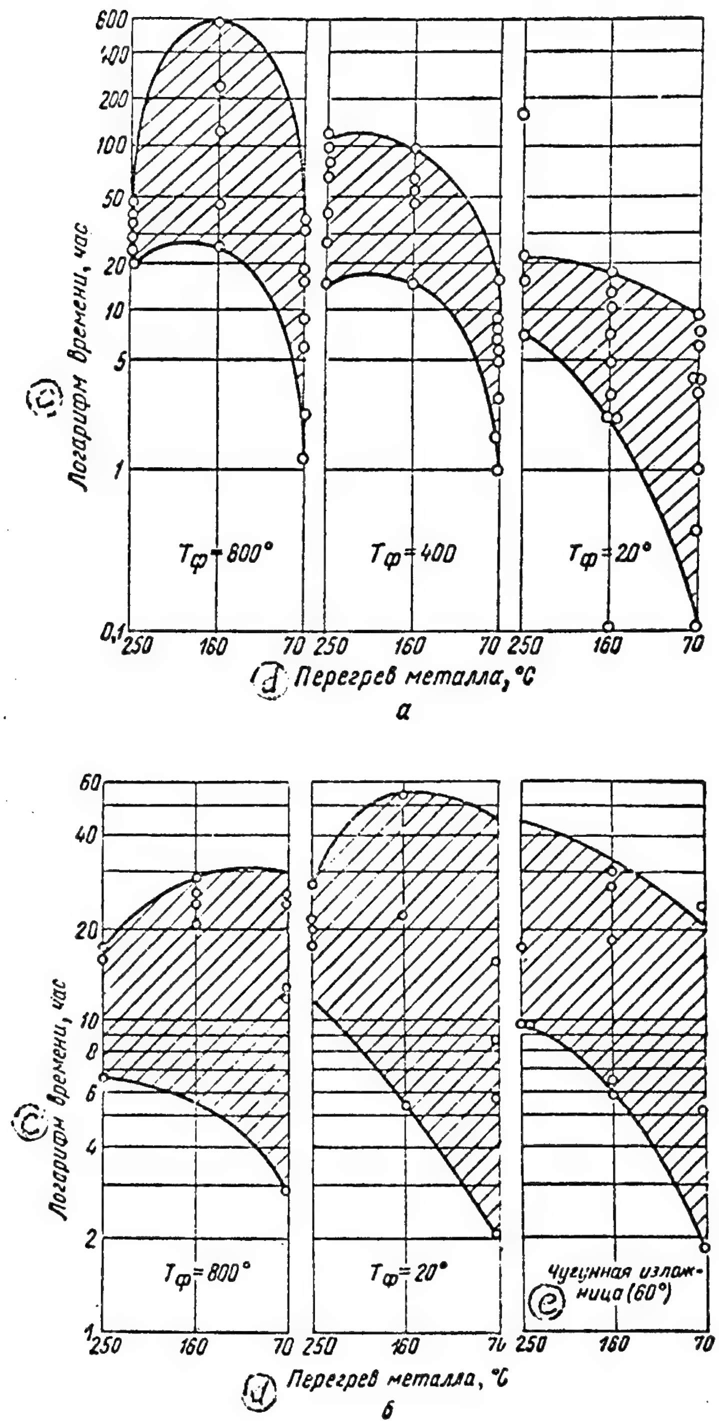

Fig 13. Stress-rupture strength of X1 steel in thin-wall (a) and cloverleaf (b) castings for σ_{τ} = 35 kg/mm^2 and T_{test} = 650C.
c - logarithm of time, hours; d - metal superheat; e - cast iron mold

of the density of the castings due to the reduction of the superheat is clearly noted only when pouring into the hot form. The impact strength of the castings produced in the cold form remains at an approximately constant level as the result of the comparatively slight changes of the structure and the density and the decisive influence of the blisters. In the cast iron form the impact strength decreases with a reduction of the superheat.

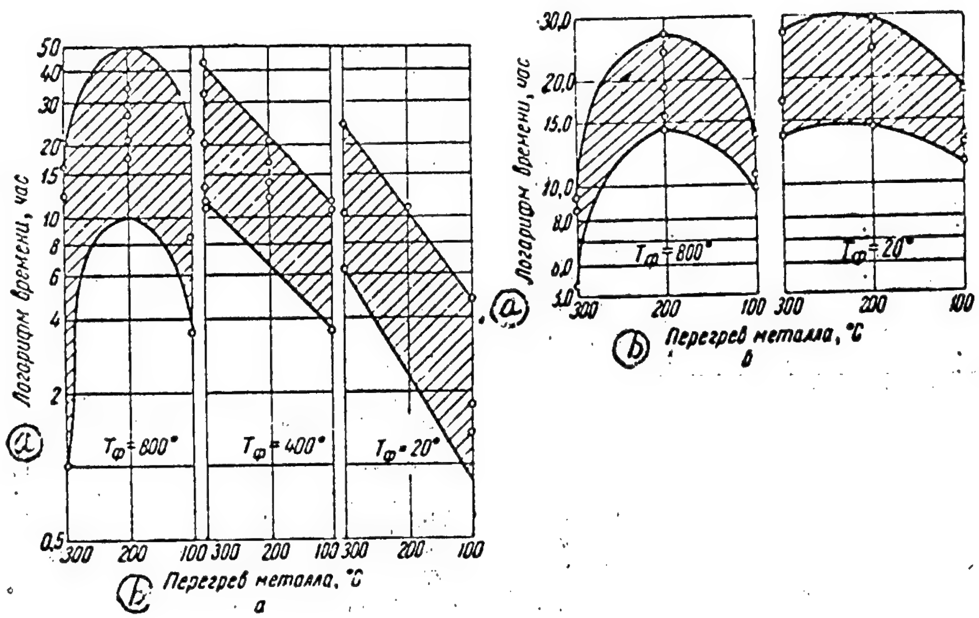

Fig. 14. Stress-rupture strength of No 6 alloy in thin-wall (a) and cloverleaf (b) castings

c - logarithm of time, hours; d - metal superheat

A good correspondence is established between the variation of the density and the stress-rupture strength of the X1 steel and the No 6 alloy (Figs 4,5,13,14) in the thin-wall and the cloverleaf castings in spite of the greater variation of the mechanical properties at the high temperature as a function of structure and in spite of the familiar scatter of the values observed in numerous cases in the testing of cast specimens.

The nonuniform distribution and the unfavorable shape of the shrinkage pores, the nonmetallic inclusions, the

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blisters, the coarse portions of the precipitation phases lead to the premature failure of the castings. Therefore, the value of the mean stress-rupture strength for given specific conditions of pouring is significantly below the individual high values for the strength of the castings with the lowest number of flaws, the most favorable shape, and the most favorable distribution through the metal of these flaws.

With a significant reduction of the rate of cooling of the castings the effect of the unfavorable structure can be greater than the effect of the density of the metal. This is explained by the clearly demonstrated reduction of the stress-rupture strength of the X1 steel and the No 6 alloy in the thin-wall castings and the significant reduction in the cloverleaf castings produced in hot forms using a high superheat of the metal. In spite of the lower density of the thin-wall castings, the stress-rupture strength of the X1 steel in the cloverleaf castings is lower as a result of the presence of coarser segments of the precipitation phases. The stress-rupture strength of the No 6 alloy, on the other hand, is higher (by about a factor of 1.5) in the cloverleaf castings thanks to the retained quite satisfactory structure, the higher density of the metal and the better conditions for the precipitation of the blisters (Fig 14 b shows the variation of the stress-yield criteria of the No 6 alloy). The mechanical properties of the No 6 alloy, whose structure is less sensitive to the cooling rate, consequently is more dependent on the density of the metal and the blisters than the X1 steel. Therefore in the usual air pouring all the factors which reduce the degradation of the castings by the blisters (high metal superheat, heating of the forms, elimination of the counterflows of the metal and air) improve the quality of the castings, their mechanical properties. In the case when the melt and the pouring of a scaling alloy is carried out in a vacuum and the danger of the degradation of the castings by the blisters is eliminated, the question of the increase of the density of the metal must be given more attention.

Results of this study indicate the logical and in general similar variation of the structure, density and mechanical properties of two alloys which are completely different in composition and class. p47

We have established the decisive influence of the casting conditions on the quality of the thin-wall casting whose properties are frequently determined not so much by the composition and the quality of the metal involved as by the general technological conditions of production (Ref 8).

The proper, theoretically substantiated, selection of the parameters of the technological process for the production of the castings is an extremely important but still inadequately utilized reserve for the improvement of the properties of the castings. Varying the casting conditions by means of the use of different form materials, the combination of heating of the form and heating of the metal, pouring methods, position of casting in the form, etc, it is possible to provide the more favorable conditions for the formation of the properties of the thin-wall castings using the highly-alloyed heat-resistant alloys.

The established patterns of the variation of the properties may sometime be disrupted as a result of the fact that each alloy has its own characteristic peculiarities. The alloys can be characterized (as a function of composition, pouring conditions and cooling conditions) by precipitation of phases of differing dispersion and differing distribution, by the presence of carbide or other liquations, by change of the gas content, by the formation of blisters or random flaws. These peculiarities can sometimes make ineffective the application of the general relations governing the formation of the structure and the properties of the castings and thus can considerably complicate the problem and hinder its direct unique solution. Therefore it is necessary to make a further compilation of experimental data on the questions covered in this study both for daily practical activity and for the development of an all-encompassing theory of the formation of the properties of the thin-wall castings.

The results of this study also make it possible to draw certain conclusions on the question of the choice of specimens for the control of the verification of the mechanical properties of the thin-wall castings. This complex problem has not to date had a final resolution since multiple and varied demands are made on the specimens for the preparation of test coupons. The specimens should be economical and simple to prepare, should provide for consistent reproduction of the values of the mechanical properties. The properties of the specimen must be comparable to the properties of the castings being verified. The most favorable conditions for the verificiation exist in the case when the specimens indicate mechanical properties equivalent to those of the castings. Even better results are obtained when testing the castings themselves under operational conditions. However, both these versions have definitely limited application: the first version is used only to determine the characteristics of isolated specific castings and the second is not

always applicable in conditions of routine production practice.

Specimens cut from the casting do not represent the structural strength of the part as a whole and can characterize only the properties of individual sections of the casting. The use of such specimens is usually limited to considerations of an engineering and economical nature.

In the selection of samples for the monitoring of the mechanical properties we must take into account the fact that each alloy has its own characteristic peculiarities. The problem of the verification of the mechanical properties is also complicated by the fact that the thickness of differing portions of the castings are not identical. Consequently, the structure, density and the mechanical properties of these portions differ. For example, the turbine blades, in addition to the thin fin, have a massive shank section. For such castings a coupon of small diameter corresponding to the thickness of the fin will not reflect the properties of the heavier shank portion of the blade. On the other hand the massive cloverleaf or wedge specimen produced in a hot form will not characterize the properties of the blade fin. These same specimens produced in a cold form might in numerous cases characterize the properties of the thin-wall castings but here it would be necessary to establish ahead of time the corresponding correction coefficients. Thus, the cloverleaf and wedge specimens which do not reflect the level of the mechanical properties of the more thin-walled castings may still be a scale factor of these properties. In comparison with precision cast specimens and the actual thin-wall castings, these coupons provide more uniform and reproducible results which is indicative of their lower sensitivity to possible degradation of the castings by various flaws. Consequently the specimens cut from massive, properly fed cloverleaf and wedge specimens characterize primarily the properties of the liquid metal and react only weakly to the deviations from the specified technological process for the production of the castings.

Precision cast specimens, on the other hand, are very sensitive to any process alteration. However this advantage of these specimens is frequently belittled without adequate basis. The presence in these specimens of various flaws which are not permissible in the actual castings (scratches, shrinkage and gaseous pits, scale, geometrical errors, presence of decarbonized layer, etc) is for some reason considered to be an inevitable evil, although these

specimens represent only a particular case of a comparatively simple thin-wall casting. But the precision cast specimens also may not reflect the properties of the thin-wall casting, they can serve only as a scale of these properties in spite of the identical thickness of the specimens and the casting and similar cooling rates. This is explained by the fact that the decisive influence on the properties of the thin-wall castings is that of the density of the metal which is in turn dependent on the feed conditions. Equivalent properties will be obtained only with identical feed conditions. However the feed conditions for the specimens and the castings are different and are determined by the positioning, the method of pouring, the constructional peculiarities of the mold-filling system, and other technological factors.

In the case of a well developed technological process for production and close control of every operation which will eliminate the danger of flaws appearing in the castings, the mechanical properties can also be verified using cloverleaf or wedge specimens. It is only necessary to take into account the arbitrary nature of this method of testing. Round and square specimens whose section dimensions approximate the dimensions of the section of the actual castings are also suitable. The choice of a particular form of specimen in the final analysis is determined by engineering and economic considerations, and the requirements for the determination of a reliable estimate of the technological properties of the alloys in the thin-wall castings.

The rate of cooling of the cloverleaf specimens cast in cold forms is significantly lower than that of the thin-wall specimens cast in the highly heated forms. From these considerations, the heating of the forms for the cloverleaf forms is not desirable since in numerous cases there may be a severe degradation of the structure and a reduction of the mechanical properties which is not characteristic of the thin-wall castings particularly in high-temperature tests. It should be noted that for the alloys having coarse grains, which are to a large degree dependent on the cooling rate, such samples may not be typical even in the case of their casting in cold forms because of the unfavorable structure.

The precision cast specimens are more sensitive and are therefore more "rigorous" specimens in comparison with the cloverleaf or other relatively massive specimens. Their advantage lies in the fact that even after mechanical work (grinding) their structural integrity is not destroyed. Therefore the mechanical testing of these specimens approaches the direct testing of the casting itself, since

they are a particular case of the casting itself. The precision cast specimens may find their application both in the field of the research and also in production operations, in the investigations of the properties of new alloys, the effect of different technological factors on the structure and the properties of the thin-wall castings, in the development of new processes for the fabrication of the mold form, methods of pouring, etc.

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EXPERIENCE IN THE PRODUCTION OF CAST TURBINE BLADES FROM HEAT-RESISTANT ALLOYS

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[Following is the translation of an article by K.P. Lebedev and M.N. Yefimova in the Russian-language publication *Trudy LPI* (Transactions of Leningrad Poly. Inst.), No 224, Moscow, 1963, pp 195-202.]

The production of cast turbine blades with precise dimensions using the various investment casting methods is of considerable interest not only from the obvious engineering and economic considerations but also from technological considerations. In the production of the cast blades it is possible to make use of alloys which are either difficult to deform or those which cannot be subjected to hot plastic working at all. It is possible to make use of complex but aerodynamically efficient profiles, hollow blades with forced cooling, and finally use can be made of blades with higher properties than is possible with stamping or rolling.

However the cast blades, particularly those used for the rotors, have been introduced into production very slowly so far. One of the reasons is the variability of the quality and the mechanical properties of the castings, but primarily it is the necessity for the testing of the rotor blades at the specified condition. (See Nekhendzi, *Steel Casting*, presented at the 23-rd International Congress of Casting Engineers in Germany, Leningrad House of Scientific and Technical Propaganda, 1958).

Below we present the results of certain tests directed toward the production of sound castings of rotor blades using various heat-resistant alloys, including alloys for moderate temperatures (up to 750C) developed in the Faculty of Casting Production of LPI.

For the investigation we selected a highly-stressed rotor blade of a gas turbine which is characteristic in shape and dimensions of a quite large group of blades whose production by casting will be of definite practical interest in the future.

Since the problem as stated did not include the resolution of problems associated with the precision (and consequently with the finishing process) of the dimensions, our work was carried out with a cast blade which corresponded exactly to the model blade (a machined forged blade) in shape and dimensions.

The wax models of the blades were fabricated on a cast babbitt press-form from a parafin-stearin 50:50 composition while the temperature-resistant coating was of hydrolyzed ethyl silicate and marshalite. We used quartz sand the filler. The removal of the model mass from the form was accomplished in a temperature chamber at 180C. The forms were baked in a PN-12 furnace at 850C. In addition to the usual free pouring of the form, centrifugal pressure was used after the form was filled with metal in order to increase the hydrostatic pressure during casting. In view of the positive effect of jet-free casting on the quality of the casting when using scale-forming alloys, we have also tried the vacuum induction method of casting developed in the casting laboratory.

Tests were conducted on the 111 austenitic alloy which contains no titanium or aluminum and therefore is not an active scale-forming alloy, on the familiar EI612 austenitic alloy which contains 1% Ti and is therefore an active scale-forming alloy, and finally on the martensitic class 15Kh11MF steel.

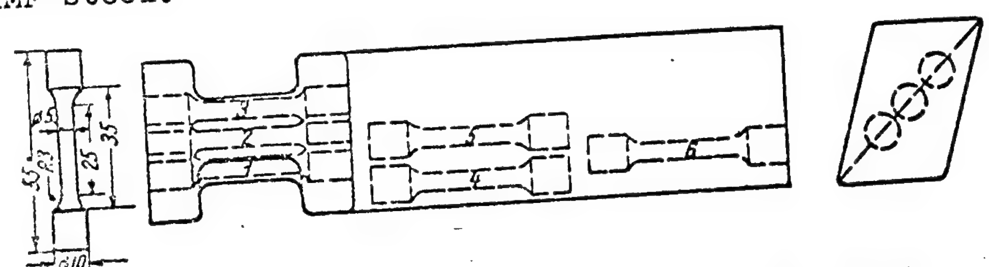

Fig 1. Pattern of cuts of specimens from the blade for mechanical testing (1-6)

Control of the quality of the cast rotor blades was exercised as follows: 1. by study of the macrostructure of coupons cut from the tips, from the thickened sections,

and from the metal inlet locations. 2. by verification of the mechanical properties by short-term tension testing of specimens cut directly from the blade in different sections (Fig 1); 3. by determination of the density of other specimens cut from the blades (Fig 2); 4. by x-ray analysis of the blades.

The requirements for the mechanical properties of the alloys tested were determined in accordance with the criteria obtained on precision cast samples prepared by the methodology developed in the casting laboratory of LPI (Table 1). (see M.T. Bogdanov, Cast Samples for Testing of Mechanical Properties of the Heat-Resistant Alloys, present collection, p 153).

Table 1

Mechanical Properties (Minimum) and Heat Treatment Conditions for the 111, EI612, and 15Kh11MF Alloys

a Марка сплава	b Режим термической обработки	σ_b кг/мм²	σ_S кг/мм²	δ, %	ψ, %	a_k (Менаже) кгм/см² c
111	f Закалка с 1250 °C, 1 час, воздух + +750 °C, 4 часа, воздух	55	25	15	20	3
ЭИ612 d	g Закалка с 1180—1200 °C, 1 час, вода + 790 °C, 10 час., воздух + 730 — —740 °C, 25 час., воздух	45	28	15	25	7
15Х11МФ e	h Отжиг при 860—880 °C, 2 часа, с печью + закалка с 1050 °C, 1 час, масло + 720—740 °C, 2 часа, с печью	70	60	10	40	4,0

a - alloy; b - heat treatment; c - Menager; d - EI612; e - 15Kh11MF; f - quench from 1250C, 1 hour, air @ 750C, 4 hours, air; g - quench from 1180-1200C, 1 hour, air @ 790C, 10 hours, air @ 730-740C, 25 hours, air; h - anneal at 860-880C, 2 hours, from furnace + quench from 1050C, 1 hour, oil @ 720-740C, 2 hours, from furnace

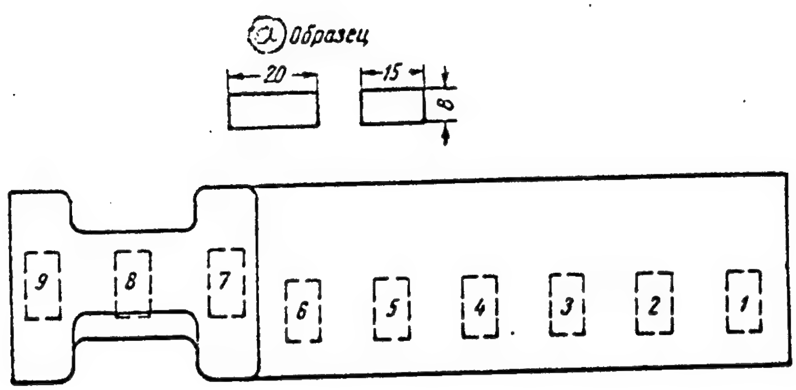

Fig 2. Pattern of specimens cut from blade for test for density

a - specimen

1. CASTING OF BLADES FROM THE 111 ALLOY

Various versions of blade casting procedures were tried in order to select a suitably simple technology, during the entire process of development of the technology, various methods were used to cast 123 blades from the 111 alloy.

The blades cast with a form temperature of 850C and a metal temperature of 1640C had good finished contour.

Consistent values of the mechanical properties and comparatively high absolute values of these properties were obtained in the production of the blades by pouring through the upper head (blade positioned vertically). For the majority of the specimens tested (73-75%) the values of the ultimate strength and the relative elongation (δ = 41-15%) correspond to the values obtained in the testing of the precision cast specimens (σ_b = 71-55 kg/mm^2).

In the siphon pouring the least consistent values of the mechanical properties were obtained in the testing of the blade fin. Particularly great spread of the data is noted in the values of the relative elongation -- only 43% of the specimens tested fell within the specification limits (δ = 26-15%). It is evident that the disruption of the directive solidification during the siphon pouring leads to

the formation of micropores which lead to a sharper decrease of the plastic properties than of the strength properties.

When pouring using the vacuum inflow method (see Nekhendzi and Bogdanov, Casting of Scaling Alloys in a Protective Atomosphere under a Vacuum, in Collection Application of Vacuum in Metallurgy, USSR Academy of Sciences Press, 1960, p 34) 75% of the specimens tested had a strength limit higher than the required value (σ_b = 69-55 kg/mm^2). However the plastic properties of the fin and the shank of the blade were not consistent -- only 35% of the specimens tested fell in the specification values for the relative elongation. In this case the poor plastic properties of the specimens are the result of the formation of micropores due to the low pressure during the induction, since during the pouring the container and the blade are in an inverted position.

The centrifugal casting method provides the highest absolute values of the strength and the most consistent strength values -- the values of the ultimate strength of 87% of the specimens tested fell in the specification range (72-55 kg/mm^2). The plastic properties were also high in both absolute value and in consistency -- the values of the relative elongation of 63% of the samples tested fell in the specification range (δ = 30-15%).

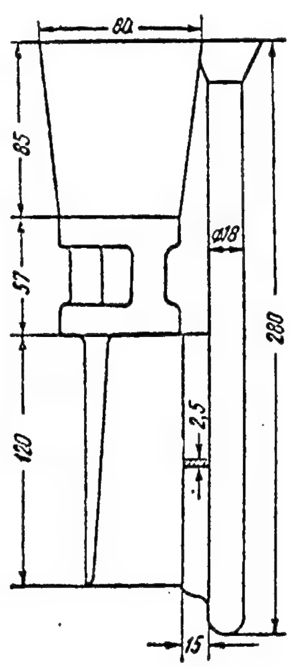

Fig 3. Diagram of blade casting. Gate system with slot feed

Of all the methods tested for the casting of the blades the optimum must be admitted to be the casting through a slotted feeder (Fig 3) which was confirmed by the following tests. After x-ray and careful visual analysis of the resulting x-ray pictures there were no defects found in the blades.

There were no defects (cracks, pores) on a macrosection taken along the thickest section of the fin and shank portions of the blade. The shrinkage flaws and porosity were completely concentrated in the header portion. Evidently the excellent impregnation of the casting is attained as a result of the directed solidification, the favorable conditions for which are created by the header system with the slot feed.

Macroetching of the blade surface discloses the varied nature of the crystallization as a function of the design of the casting block. In the macroetching of the fin of blades produced by pouring through the upper head, there are clearly seen three zones of crystallization: fine-grained, transcrystallization, and a zone of randomly oriented grains. However with pouring through the slotted feeder, large randomly oriented grains are seen over the entire surface of the fin (Fig 4). Evidently the uniformity of the structure in the latter case must also provide for the uniformity of the mechanical properties. On the macrosection of the fin of such a blade there is found a single-phase structure -- austenite. The macrosection made along the thickest section of the same blade indicated the complete absence of shrinkage flaws and cracks along the grain boundaries.

9 coupons were cut from each of the blades for the determination of the density. The dimensions and the location of some of the coupons are shown in Fig 2. The density of the coupons was determined using the method of hydrostatic weighing in toluene.

From the data obtained we see that the density varies only slightly along the height of the blade (Fig 5). The difference between the maximum and minimum values amounts to 0.2% which is indicative of the uniform impregnation along the entire height of the blade. The absolute values of the density of the specimens cut from the blade of the 111 alloy vary in the range 8.000-8.015 g/cm^3 and differ little from the precision cast specimens (8.020-8.087 g/cm^3). The high absolute value of the density of the samples cut from the blades is indicative of the excellent impregnation of the castings along the length using this particular

casting technique.

From the results of the mechanical test, 76% of the samples fell in the specified density range (73% fell in the relative elongation limits). It is important to note (Table 2) that the most consistent values are those of the mechanical properties of the fin of the blades, particularly the values of the ultimate strength. During the testing of the shank the deviations are primarily associated with the central portion of the shank and are evidenced primarily in the reduction of the plastic properties. The absolute values of both the strength and the plasticity are adequately high. The production of the blades by casting by means of the header system with slotted feed provides for smooth filling of the form, the possibility of the directed solidification and excellent impregnation of the of the casting, uniformity of the structure and properties along the entire blade and consistent mechanical properties.

Table 2

Mechanical Properties of Specimens from Blades Produced by Casting Through a Slotted Feeder

№ плавки (a)	(b) Номер образца											
	1		2		3		4		5		6	
	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %
18	56,2	22,4	45,8	8,0	62,3	21,8	62,3	18,7	62,7	23,1	53,2	12,5
18	56,6	15,3	52,0	12,0	62,3	16,3	—	—	56,2	21,7	57,2	23,4
18	48,0	10,4	46,4	6,9	48,5	12,6	60,2	21,0	60,7	24,5	53,6	15,5
19	60,2	18,5	56,2	13,2	56,2	17,6	59,4	14,1	59,4	22,9	59,3	20,0
19	56,2	21,2	57,7	15,3	—	—	57,1	16,2	59,7	17,8	59,7	17,8
19	63,4	18,3	39,2	6,3	64,3	18,4	57,2	19,7	60,2	20,0	58,1	23,0

a - melt No; b - specimen number

2. CASTING OF BLADES FROM THE EI612 ALLOY

In the casting of blades from the EI612 alloy we observed scale formation, which hindered the filling of the form, particularly of the thin sections. In external appearance these blades produced by casting through a slotted

feeder had a less clearly finished contour of the thin edge of the fin than the blades of the 111 alloy produced by the same method. Excellent blade contour was obtained when casting by the methods of vacuum induction and with the use of centrifugal pressure.

A comparison of the mechanical properties of the samples from the blades produced from the EI612 alloy by three different casting methods (Table 3) indicates that the highest strength properties in both absolute value and in consistency is provided by casting through the slotted feeder (87% of the samples tested fell in the specified range). The lowest absolute properties, in spite of the good consistency (81%), were found in casting with the use of centrifugal pressure.

The plastic properties of the samples, regardless of the casting method, are very low. Values of the relative elongation less than 10% are shown by 87% of the samples using slotted pouring, 96% using vacuum induction, and 80% using centrifugal casting. In the vacuum induction and the centrifugal method there are no obstructions to the passage of scale. Therefore the scale forms in a large amount and reduces the strength properties of the metal.

The lower strength properties of the blades produced by the centrifugal casting method are explained by the fact that during the vacuum induction the scale is brought into the form and after rotation of the form into the normal position the scale floats upward only partially. The separation of the scale under the influence of the centrifugal force, toward the center of rotation, cannot take place because of the rapid solidification of the metal and the scale remains in the casting. In this method of casting, in comparison with other methods, the scale noticeably reduces the strength properties but it reduced the plastic properties to an even greater extent. The slight improvement of the plastic properties of the blades produced by centrifugal casting in comparison with the other methods (Table 3) is explained, apparently, by the higher degree of compaction of the metal under the action of the centrifugal force. But still the improvement in the plasticity in this case (because of the presence of the scale) is slight.

Thus, the best properties of the blades cast from the EI612 alloy can be obtained by filling the form through a header system with slot feeder.

Table 3

Mechanical Properties of Samples from Blades Cast from EI612 Alloy by Various Methods

Способ заливки (a)	№ плавки (c)	(b) Номер образца											
		1		2		3		4		5		6	
		σ_b, кг/мм²	δ, %	σ_b, кг/мм²	δ, %	σ_b, кг/мм²	δ, %	σ_b, кг/мм²	δ, %	σ_b, кг/мм²	δ, %	σ_b, кг/мм²	δ, %
Через щелевой питатель (d)	20	63,4	8,6	58,6	3,3	62,7	10,0	53,2	3,3	56,2	4,9	51,5	2,5
	21	57,2	12,6	40,8	0,8	66,5	12,7	44,8	2,5	60,6	9,8	52,2	4,0
	22	52,6	2,6	63,0	4,4	69,0	6,7	56,2	2,2	61,4	7,1	53,6	1,3
	22	43,0	0,4	56,3	4,4	56,2	2,3	61,4	3,1	64,8	6,3	56,3	1,7
Вакуумный засос (e)	20	44,8	1,0	34,2	0,6	53,6	2,2	47,0	0,8	60,2	8,7	45,8	1,0
	20	45,8	1,9	43,4	0,8	55,2	5,5	40,8	1,1	43,8	0,5	45,3	0,6
	21	58,7	4,6	45,5	0,6	56,2	7,9	54,5	2,5	59,7	5,3	56,6	3,4
	21	53,2	10,5	39,8	0,6	59,3	4,1	58,7	6,2	61,4	5,5	52,2	3,2
Центробежный способ (f)	23	42,8	8,1	42,3	9,1	55,0	17,5	47,0	6,1	53,8	12,2	53,6	10,7
	23	50,0	15,3	45,8	15,2	—	—	—	—	56,2	15,3	52,0	9,2
	24	45,4	14,0	49,5	11,2	—	—	49,0	7,5	56,2	9,3	45,8	6,7
	24	45,8	11,4	49,7	7,9	44,7	12,3	50,0	9,9	52,1	7,3	43,8	2,7

a - casting method; b - specimen number; c - melt number; d - through slotted feeder; e - vacuum induction; f - centrifugal method

3. CASTING OF BLADES FROM THE 15Kh11MF ALLOY

In the casting of blades from the 15Kh11MF alloy there is also observed some scale formation but to a much lesser degree than when casting of the blades from the EI612 alloy. Therefore in external appearance the blades have a well finished contour.

From Table 4 we see that the strength properties of the blades have excellent absolute values and high consistency -- 88% of the samples tested fell in the specification range.

Table 4

Mechanical Properties of Samples from Blades Produced from the 15Kh11MF Alloy by Pouring through a Slotted Feeder

№ плавки	Номер образца											
	1		2		3		4		5		6	
	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %	σ_b кг/мм²	δ, %
25	83,1	16,3	80,1	8,5	82,0	14,5	86,1	4,9	36,7	0,4	82,0	2,2
27	83,1	13,7	81,2	5,6	85,2	13,0	80,5	10,6	87,3	10,5	81,6	3,4
27	80,6	12,3	81,2	8,4	81,6	12,6	87,4	11,2	81,6	4,5	85,6	5,1
27	79,2	13,5	83,1	9,1	82,0	14,1	82,5	12,9	79,6	2,1	83,5	6,3
27	79,8	12,4	82,6	13,0	78,6	12,7	84,6	6,2	88,2	10,3	82,0	5,7
28	83,7	13,8	62,2	0	82,6	13,7	88,2	7,8	83,0	5,3	72,5	0,6
28	82,5	15,0	69,3	1,0	82,6	16,6	85,6	4,5	67,3	0,9	72,0	1,0
28	82,5	13,3	81,5	3,6	82,1	15,1	74,6	2,4	79,7	11,2	78,5	8,7
28	84,6	13,7	70,0	1,1	84,5	14,6	82,5	5,1	83,6	2,5	60,8	0,6
28	84,2	3,6	86,2	14,3	85,5	12,8	87,4	7,8	69,4	2,6	60,2	1,6

a - melt No; b - specimen number

CONCLUSIONS

1. The highly stressed cast turbine rotor blade which is characteristic of a large group of blades can be produced by the free casting method of investment casting in the case where the alloy is not scale-forming. Here the consistent set of mechanical properties is obtained by pouring through a header system with a slotted feeder.

2. The non-scaling 111 alloy developed in LPI which has comparatively good casting properties provides in all properties sufficiently high operational qualities of the cast turbine rotor blades.

3. In the casting of the blade from the EI612 and 15Kh11MF alloys, in spite of the high values of the ultimate limit and the fluidity, in the majority of the cases the elongation is severely reduced both in atmospheric centrifugal casting and when using vacuum induction. This takes place as a result of the intense scale formation (of the EI612 alloys) and the nonuniform distribution of

the ferrite (15Kh11MF) alloys).

4. Successful application of the scaling type EI612 alloys for the casting of blades can be achieved by the use of those casting methods which provide for the absence of the scale in the melting and pouring process with sufficient assurance. To do this we must use either vacuum melting and pouring or make further investigations of the development of the metal supply system for the vacuum induction.

5. Uniform distribution of the ferrite during the casting of blades from the 15Kh11MF alloy can be achieved by an optimum concentration in the alloy of C, Cr, Si and Ni for a given rate of cooling using a given technological process and blade dimensions.

6. The centrifugal casting method can significantly improve the conditions of the feeding of the casting and consequently can provide for consistent quality. The practical use of this method for the scale-forming alloys will require the use of a protective atmosphere.

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